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WADC TECHNICAL REPORT 54-175
PART 1

NOTCH SENSITIVITY OF HEAT-RESISTANT ALLOYS AT ELEVATED TEMPERATURES

Part 1. Preliminary Studies of the Influence of Relaxation and Metallurgical Variables

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UNIVERSITY OF MICHIGAN

AUGUST 1954

WRIGHT AIR DEVELOPMENT CENTER

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August 1954

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FOREWORD

This report was prepared by the University of Michigan, under USAF Contract No. AF 18(600)-62. The contract was initiated under Research and Development Order No. 614-13 MC, "Design and Evaluation Data for Structural Metals", and was administered under the direction of the Materials Laboratory, Directorate of Research, Wright Air Development Center, with Dr. A. Herzog acting as project engineer.

ABSTRACT

Tests have been performed seeking to understand the factors affecting notch sensitivity of heat-resistant alloys under sustained loads at elevated temperatures. The investigation was based on the belief that varied response to notches must be related to relaxation characteristics of alloys at service temperature. A material was postulated to be strengthened or weakened by a notch according to the portion of total rupture life consumed while initial stress concentrations around a notch are reduced and redistributed by a creep-relaxation process. A procedure was proposed whereby the history of representative fibers in a notched specimen would be followed to the point of rupture.

Data from other sources comparing strengths for smooth and notched bars of materials of interest are included. Additional data required for the proposed analysis were obtained under the present program for three alloys with conventional heat treatments:

S-816 at 1350°F
Waspaloy at 1500°F
Inconel X-550 at 1350°F

Test results included stress - rupture time properties, short-time tensile properties, and creep properties when stresses were changed from one level to another during a test. Relaxation characteristics were measured for initial stresses both below and above the proportional limit.

The notch strengthening observed for S-816 and Waspaloy, and the notch weakening for Inconel X-550 at the test temperatures has been satisfactorily explained in terms of comparative relaxation and stress-rupture time characteristics, though further work is indicated before a quantitative correlation is attempted.

Tests were conducted to determine the effect of some metallurgical variables on the notched bar rupture test characteristics. Cold working had the greatest effect on notch sensitivity of the several conditions investigated, but no severe case of notch weakening was observed for either S-816 at 1350° and 1500°F, or for Waspaloy at 1500°F in the limited number of tests.

PUBLICATION REVIEW

This report has been reviewed and is approved.

FOR THE COMMANDER:

M. R. WHITMORE
Technical Director
Materials Laboratory
Directorate of Research

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INTRODUCTION

Engineers are constantly faced with the problem of allowing for effects of stress concentrations in structures they design and build. Relief of local high stresses is associated with ability of a material to redistribute stresses to a more favorable state before the fracture strength of the material is exceeded. At high temperatures, where creep occurs during service, reduction of stress concentrations by relaxation would appear to be an important factor in this process of stress redistribution.

Increasing use has been made by engineers of rupture tests conducted on notched specimens to evaluate ability of alloys to withstand stress concentrations in prolonged high-temperature service. In such tests, some materials have been found to be strengthened by notches; others may be weakened. A third group shows both strengthening and weakening, depending on the length of the test; i.e., on the stress level employed. A common case is for an alloy to be strengthened by the presence of a notch in short-time tests, but to be weakened for long-time runs at low stress levels. A few instances have been reported of recovery from notch weakening at prolonged time periods.

This investigation has been based on the belief that observed variation in response to notches for different materials, or for the same material at different temperatures, or for differing notch geometry, must be closely related to stress-time relationships as controlled by creep relaxation. Favorable redistribution of stresses through the relaxation process reduces the effective stress either rapidly or slowly, depending on creep characteristics of the particular alloy at the test temperature. If the residual effective stress drops quickly enough, life is increased by presence of the notch; if relaxation is slow, a major portion of the total life is quickly expended in areas of initial high stress. In the latter case, residual stresses around the notch may eventually fall to quite low values, but by this time only a small portion of rupture life remains and failure occurs earlier than for a comparable smooth bar.

At the outset it may be reasoned that a notch introduces nothing inherently new into properties of an alloy, but only changes the stress-strain histories of fibers in the notched bar. If one were to reproduce in a smooth bar the history experienced by a fiber of a notched bar, the life of each should be the same. It is not sought here to explain why a smooth-bar rupture specimen behaves as it does. Rather, it is hoped that, given the properties of the alloy as ordinarily tested in tension, one might extend these results to other cases where stress conditions are not uniform, but where the stress distribution may be estimated.

Analysis of the Problem

A circumferential notch in a tensile specimen introduces at the root of the notch a high axial stress (S_a) and a hoop stress (S_h) which is smaller, but still higher than the nominal stress. (By nominal stress is meant the axial load divided by the minimum cross section of the bar in the plane of the notch root.) A radial stress (S_r) is also created, starting with a value of zero at the root of the notch and always remaining smaller than the nominal axial stress for radii nearer the axis of the specimen. So long as the yield point of any fiber is not exceeded, the distribution of stress in a notched bar may be closely approximated using Neuber's theoretical analysis for a deep hyperboloid (Ref. 1, Chap. V). Beyond the yield point, exact stress and strain distribution in regions of plastic flow must be obtained experimentally.

When stresses in different directions are present in a material, they interact. Forces in different directions tend to cancel each other's effect when both are tensile or both are compressive. According to the maximum shear-strain energy theory of yielding, for ductile materials the stress ($\overline{S}$) which is effective in causing plastic flow to take place may be computed from the expression

$$2\,\overline{S}^2 = (S_a - S_h)^2 + (S_a - S_r)^2 + (S_h - S_r)^2 .$$

For elastic conditions the effective stress over part of the bar will still exceed the nominal because Sa is so high. The maximum stress concentration occurs at the root of the notch. For fibers nearer the axis of the bar, each of the stresses becomes smaller, differences between pairs of stresses become even smaller, and the effective stress for these fibers becomes a rather small fraction of the nominal stress.

The effective stress defined above has been verified as a satisfactory combination of the individual stresses in correlations of yielding and plastic flow of ductile metals at room temperature and of creep of metals at elevated temperatures when constant multiple stresses are acting. It appears reasonable to use this same effective stress in the first attempts to correlate rate of relaxation of variable multi-axial stresses.

The variable effective stress existing during life of a fiber will be tentatively assumed the significant factor determining time until it ruptures. The general method of analysis may still be adapted to any better combination of principal stresses suggested at a later date. For a notched bar to have a longer rupture life under given nominal axial stress than does a smooth bar, the assumption is that the notch reduces the effective stress below the nominal value during a major portion of the test. It appears that this could occur by one or more of the following ways:

1. Plastic yielding upon application of the load.

2. Progressive change in shape of the notch by the normal creep process.

3. At elevated temperatures, stresses present in restrained portions of metal tend to decrease with time by a process of stress relaxation. Elastic strains are replaced by creep deformation without significant change in gross dimensions of the part. (This is similar to behavior of a tightened bolt in an unyielding flange at elevated temperatures, with gradual drop in the pull exerted by the bolt while its stretched length remains fixed.)

The rate of such relaxation is non-linear so that the higher the initial stress, the faster it falls. It is to be expected that extreme stress concentrations found at the root of a sharp notch on loading should relax rapidly. The variation in rates of relaxation for the three principal stresses of a fiber should cause even more rapid reduction in the differences between these principal stresses. And it is these stress differences which determine the effective stress.

For the presence of a notch to shorten rupture life, the combination of stated mechanisms apparently does not reduce the effective stress below the nominal stress early enough in the test. Stated in another way, too much of the total life has been used up while getting down to a low effective stress.

Besides the stresses which are present, the metallurgical condition of an alloy should affect response to stress concentration. As an example, localized yielding may occur near the notch root. Such plastic working may either strengthen or weaken affected fibers depending on the particular alloy and its initial condition, on the temperature, and on the length of time the test runs before failure. Working can also change the amount of deformation which occurs before fracture initiates in a fiber of a metal. It is quite possible that local yielding at the start of a test may alter creep and relaxation properties.

Any prior metallurgical treatment which changes the proportional limit and work-hardening characteristics, the material's ductility, its creep rates, or its rupture life under constant load should be reflected in the relative behavior of notched versus unnotched bars under stress-rupture conditions. In particular, low elongation of smooth bars at rupture is known often to be associated with notch brittleness under high-temperature conditions.

Geometry of the notched specimen, especially the sharpness of the notch root, is also known to affect notch-bar rupture strengths. But any notch geometry for which the starting stresses and strains can be determined should prove equally amenable to the general analysis proposed above.

SECTION I

COMPILATION OF PRINCIPLES AND DATA FROM NOTCHED-BAR RUPTURE TESTS

The first step in this investigation was to accumulate and analyze available data on notched-bar rupture tests for alloys of the types used in aircraft gas turbines. The compiled data, Figures 1 through 13, show the following results of interest.

1. For rupture times up to 1000 hours, the data showed S-816 to be notch ductile at 1200°, 1350° and 1500°F. (See Figure 1.) Figure 2 indicates that the same conclusion holds for S-816 (Cb + Ta) alloy. The data fail to show a conclusive influence of variation in solution temperature or in notch configuration.

2. M252 alloy was strengthened by the presence of the notch (Figure 3) when given the standard heat treatment for this alloy.

3. Very limited data on Inconel X showed greater sensitivity for a bar made by grinding flats on opposite sides of a notched cylindrical bar than did the original round bar with the same notch. (See Figure 4.) This result might have been anticipated from the higher elastic stress concentration factor for a flat bar with a given notch shape compared with a three-dimensional specimen.

Where data were available for notches with 0.005-inch root radius and also for a sharp notch ($r \leq 0.002$ inch), the sharp notch had the lower strength. At 1350°F the sharper of the two notches exhibited notch weakening, while the bar with 0.005-inch root radius was strengthened by the notch.

4. Davis and Manjoine (Ref. 6) treated Refractaloy 26 so as to obtain rather different diamond pyramid hardness at constant small grain size and to give coarse versus fine ASTM grain size at roughly the same hardness levels. A wide range of root radii was employed for constant bar diameter, notch depth and notch angle. Curves for several intermediate values of r/d have been omitted in plotting the data on Figure 5.

For all treatments it was found possible to produce bars with both higher and lower strength than for smooth bars by suitable choice of root radius alone. Moreover, onset of notch sensitivity was not a function of any single universal value for ductility of the unnotched bars. For $r/d = 0.2$, the fine-grained alloy with 330 DPH number was strengthened for rupture times where smooth-bar elongations were about 7 percent. At like times to failure, the alloy in the coarsened condition had better unnotched ductility, but was definitely notch brittle.

5. Results for specimens of K-42-B (42% Ni, 18% Cr, 22% Co, 2.2% Ti, 0.2% Al, 0.05% C, Bal. Fe) with constant composition but different prior heat treatments are compared in Figure 6. For both conditions rather severe notch sensitivity was apparent at all stresses tested. Smooth-bar elongations were very low (0.4 to 1.9 percent). Of the two treatments investigated, the one accompanied by lower elongations at rupture gave the higher rupture strengths for notched and smooth bars alike.

6. Three degress of cold working of 16-25-6 (Timken Alloy) after forging show little effect on strength of notched bars despite a pronounced rise in rupture strength upon cold working. (See Figures 7 and 8.)

7. A memorandum from the Thomson Laboratory of the General Electric Company (Ref. 7) gave data on notch sensitivity of specimens taken from a number of 16-25-6 turbine-wheel-rim forgings made by conventional hammer cold working and by a die-expansion process. No correlation from rim to rim was evident between notch brittleness and the elongation at rupture of smooth bars. (See Figures 9 through 11.)

A few specimens, both smooth and notched, cut in the tangential direction from die-expanded rims had consistently longer rupture lives than those samples in the radial directions. These specimens showed elongations at rupture covering a wide range from 1.7 to 30 percent.

Life was on the high side for all bars from a single rim forged by conventional hammer cold working after prior solution treatment at 2000°F, instead of the usual 2100°F (Figure 12). In contrast, solution treatment at the lower temperature before die expanding appeared to result in weaker specimens.

When specimens from die-expanded rims were aged for increasing lengths of time at 1200° and 1350°F, radial specimens appeared to vary more with aging than did bars cut in the tangential direction. In general, after an initial strengthening, further aging served to give first a decrease and then an increase in time to rupture for both smooth and notched bars (Figure 13).

Conclusions Drawn from Compiled Notch Data

The most evident conclusion is that notch sensitivity is a complex function of notch geometry, physical properties and microstructure. The large influence of notch geometry makes it difficult to generalize other factors involved in notch-bar testing. The following summaries appear to be pertinent.

For any given material it has been demonstrated that there is at least a qualitative relationship between elastic stress concentration factor and strength where variation in the factor is obtained by changing the root radius of the notch. When the notch depth was varied at constant root radius under conditions giving nearly constant stress concentration factor, however, variations in effects on strength were obtained.

As the notch increases from a very dull to a very sharp notch, the strength increases to a maximum and then decreases to a point where rupture time is reduced more and more as the sharpness increases. The notch radius at

which strength starts to decline is a function of both ductility and metallurgical characteristics. (See Ref. 6.)

Ductility and Notch Effects

In those cases where ductility changes with time for fracture, it appears that notch-sensitivity effects will qualitatively follow the ductility changes. According to Brown and co-workers (Refs. 5 and 8), these effects can occur at a rather wide variation in ductilities. Data gathered in this report support that finding. Materials having elongations in the range from 2 to 15 percent in the rupture test may or may not show notch sensitivity, depending on notch geometry and metallurgical characteristics. For commonly-used notches, alloys with elongations above this range should be notch ductile; those with less elongation will normally be notch brittle. Particularly interesting was Brown's observation that for sharp notches sensitivity could develop at quite high ductilities when structural changes occurred during tests.

Brown and his associates have presented data for rupture of notched and smooth bars where the alloy under study is notch sensitive only over a limited range of stress. For the materials studied, it was found that elongation of the smooth bars at rupture seems to be related to the presence or absence of notch sensitivity.

For short times to rupture where the elongation values are high, the notched bar was stronger. As elongation fell off the notched-bar strength dropped below that of the smooth bars, only to rise above it again when the elongation increased at very long testing times.

Metallurgical Factors

It is practically impossible to generalize regarding metallurgical factors. Apparently for a given solution treatment, variation of the ductility by changes in aging conditions will alter the strengthening or weakening effect of a given notch. In general, decreasing the ductility tends to increase notch sensitivity. If the solution treatment is changed, then the effects of aging may be different.

Ductility alone appeared to be an inadequate measure of the effect of a change in structure. Available results show that change in grain size alone is also not a perfect indicator.

Viewed as a whole, the data confirm the belief that commercial alloys now used in jet engine components are not sensitive to notches when they respond normally to recommended conventional treatments. However, with a critical combination of deviations from conventional practice it is possible to get marked deterioration from desired properties.

SECTION II

SOME EXPERIMENTAL HIGH-TEMPERATURE PROPERTIES AND THEIR INTER-RELATIONS

Several different types of experimental data appeared to be needed to relate creep relaxation to notch properties, and to separate effects of plastic yielding on loading and of relaxation of stresses originally concentrated near the notch root:

1. During application of the load, an alloy must possess a suitable combination of yield point and ductility to prevent stress concentrations from exceeding the tensile strength. Ability to deform and relieve high stresses is associated with the shape of the stress-strain curve.

Any plastic deformation which does occur on loading will affect the state of stress distribution used as a starting point in the proposed calculations seeking to relate notch effects and relaxation characteristics. Analysis of this initial condition required knowledge of plastic strains around the notch, together with stress-strain curves for each alloy at each temperature considered.

Experiments on change of notch shape upon loading were not included because such tests are being performed by another contractor working with these same alloys and test conditions.

2. Assuming that a stress concentration will remain after the load is applied, then the time to rupture should depend on the rate at which this stress concentration is reduced by replacing elastic strains with plastic creep strains. Relaxation properties were needed for the range of stresses expected to be found in typical notched bars. Tests were also needed to determine how such relaxation characteristics are affected by plastic strain occurring during loading.

3. It has been proposed to follow changes in representative fibers as creep and relaxation occur. Rupture life of the fibers under consideration is then to be compared with life for other fibers subjected to a like history in a smooth bar. Before any such analysis can be completed, it is necessary to know how varying the applied stress influences total time for rupture, i.e., how to add fractions of rupture life under different stresses.

A suitable answer was sought by conducting tests with smooth bars for a constant stress over part of the run and then changing to a higher or lower constant stress for another portion of the test.

4. Material properties of high-temperature alloys may alter during service. Any factor which changes the physical properties or rupture life of fibers in a smooth bar should also affect fibers in notched bars. Consequently, consideration had to be given to the effects of such metallurgical changes under the conditions existing at the root of the notch. It was recognized that these could be quite different from those normally considered in smooth-bar tests due to initial plastic flow on loading or while the stress concentration was at a high level.

Materials Tested

Materials tested included three typical metallurgical types of turbine blade alloys used in jet engines: a cobalt-base forging alloy (S-816), a chromium-nickel alloy precipitation hardened with Cb, Ti, and Al (Inconel X-550), and a chromium-nickel-cobalt-molybdenum alloy precipitation hardened with Ti and Al (Waspaloy).

Chemical composition of the alloys, in weight percent, was furnished by the suppliers:

Element	S-816 (Heat 63730)	Waspaloy (Heat 44036)	Inconel X-550 (Heat Y-7180-X)
C	0.38	0.08	0.05
Mn	1.22	0.80	0.73
Si	0.49	0.61	0.28
P	0.012	0.017	--
S	0.018	0.017	0.007
Cr	20.04	18.72	14.92
Ni	19.43	Bal	Bal
Mo	3.98	2.93	--
W	3.93	--	--
Cb	2.89	--	1.03*
Co	43.32	13.44	--
Fe	3.44	1.17	6.59
Al	--	1.29	1.16
Ta	0.85	--	--
Ti	--	2.29	2.5
Cu	--	0.10	0.03

* Cb + Ta

Specimens were given the following conventional heat treatments in University of Michigan laboratories prior to machining:

S-816	Waspaloy	Inconel X-550
2150°F, 1 hr, W.Q. +	1975°F, 4 hrs, A.C. +	2150°F, 1 hr, A.C. +
1400°F, 12 hrs, A.C.	1550°F, 4 hrs, A.C. +	1600°F, 4 hrs, A.C. +
	1400°F, 4 hrs, A.C.	1350°F, 4 hrs, A.C.

Typical microstructures after heat treatment are shown in Figures 14, 15, and 16.

All specimens were turned between centers on a lathe and then hand polished.

Test Conditions

Since the object of the program was to study factors affecting notch sensitivity, it was attempted to choose test conditions which would include one notch ductile alloy (S-816 at 1350°F), and one alloy in a notch brittle condition, at least for very sharp notches (Inconel X-550 at 1350°F). For a third condition (Waspaloy at 1500°F), it was hoped to obtain a borderline case without decided notch strengthening or weakening.

Carlson and Simmons (Ref. 9) have since reported comparative rupture lives of smooth and notched bars from the same heats of alloys as used in the present investigation. Notched bars used had three circumferential notches, each with different root radius. Results of completed tests have been reproduced in Figures 17 through 19. In these plots the notch-bar points are distinguished according to the root-radius of the notch which failed. The data show that test temperatures chosen for the present program gave two cases of notch strengthening (S-816 at 1350°F, Waspaloy at 1500°F). The third condition (Inconel X-550 at 1350°F) showed slight notch strengthening in short-time tests, with progressive notch weakening at lower test stresses.

All tests for the present investigation were performed in individual beam-loaded creep-rupture units. Axial deformation was measured by a Martens-type optical extensometer with a sensitivity of from 3 to 4 x 10^{-6} inches/inch of strain. Temperatures were controlled to ± 3°F.

Usual practice was to place a specimen into a cold furnace the night before the test was started and to alloy the furnace to rise to 100°F below the desired testing temperatures. The specimen was brought to the final value and temperature distribution adjusted over a period of between 2 and 5 hours just prior to loading.

Tensile Properties

Short-time tensile characteristics for all three conditions studied are shown on Figure 20. The following properties were obtained:

Property	S-816 at 1350°F	Waspaloy at 1500°F	Inconel X-550 at 1350°F
Yield Strength (0.2 percent offset--psi	47,500	76,500	89,500
Tensile Strength--psi	83,500	81,000	104,000
Proportional Limit--psi	36,000	48,000	72,000
Elongation--percent/2 inches	39	6	4
Reduction of Area--percent	33.5	7	7
Elastic Modulus--psi	23×10^6	21×10^6	23×10^6

Relaxation Characteristics

The term relaxation has been used by metallurgists in at least two contexts. Here it is used to mean replacement of elastic strains by plastic strain in the form of creep. Relaxation characteristics are commonly plotted as a smooth curve of residual stress versus time for a given initial stress and for continued reduction of load so as to maintain the total strain constant. In tests the stress reduction is often performed in finite steps (see Figure 21). Such a procedure was used in this investigation. The specimen is loaded to its highest values (S_1 at point A) and the strain measured. Creep is then allowed to occur (A-B) until such a time that removal of a weight will return the specimen to its original length (point C), but at a lower stress (S_2). When a large number of small equal weights are used, the resulting step-wise curve of residual stress versus elapsed time approaches the theoretical smooth curve.

There is a practical limit of weight decrement below which it is not advisable to go. Accuracy is greatly reduced if the length of time for a relaxation step to occur is of the same order as that for unavoidable small temperature cycles; or if the strain decrement upon removal of a load is not considerably larger than the sensitivity of the extensometer system. (A change of temperature of 1°F results in a change in length of approximately 10×10^{-6} inches/inch, which is some two or three times the sensitivity of the extensometer used.)

With a few exceptions, relaxation data published in the past have been run at stresses somewhat below the proportional limit. (See Ref. 10.) At such relatively-low stresses the relaxation process is slow enough to permit accurate determination of the time when a weight should be removed in the step-down type of test. In the present investigation the higher stresses permit less exact determination of the proper time for relieval. The matter of reproducability of results during the early part of a test becomes of concern.

Results from duplicate runs with Inconel X-550 and Waspaloy are compared in Figure 22. Also shown are data for three specimens of S-816, all relaxed at 1350°F from the initial stress of 40,000 psi. For each material the scatter between tests was quite small compared with the large drop in stress which occurred by relaxation.

Relaxation characteristics are shown separately for the three materials in Figures 23 through 25. All three plots are presented to the same scale

for easy comparison with one another and with Figure 22. Time has been plotted on a hyperbolic sine scale, rather than the usual log scale, so that the early portion of the curves could be included all the way to zero time. At times up to one hour, this scale is nearly linear, allowing easy interpolation of relaxation curves in the region of greatest interest in the present study. At times beyond 10 hours, the hyperbolic sine scale and the logarithmic are essentially the same.

When the sets of relaxation curves are compared, both contrasts and similarities are apparent. In all three cases, the residual stress level after a period of less than 100 hours was nearly the same for a given alloy and temperature, regardless of the starting stress. At the end of 20 hours, all Inconel X-550 specimens were at about 45,000 psi, while all tests for the other two alloys had residual stresses of one-third this value or less after the same time interval.

The most significant difference between the sets of data is the short rupture life of the Inconel X-550 at the 45,000 psi residual stress (about 260 hours), contrasted with several thousand hours for Waspaloy and S-816 at a residual stress of 15,000 psi at their respective test temperatures.

In all three cases studied, for the lower stresses the rate of decline of stress level was moderate and the individual curves slowly approached common values over a period of from 10 to 100 hours. This behavior appears to be typical for all stresses below the proportional limits measured at the test temperature. Higher starting stresses exhibited an abrupt drop in stress level, with relaxation curves for a high initial stress crossing over those started at lower values. In a matter of hours the residual stresses were nearly in exact opposite order to those at the start of the relaxation tests for the conditions investigated. Although this was somewhat unexpected behavior, there appears to be sufficient substantiating relaxation data in the literature to indicate that plastic deformation on loading does reduce resistance to relaxation.

The questions remain as to the effect on relaxation behavior of rapid straining such as might occur in fibers near the root of a notch when the load is applied to a notched tensile specimen. An indication of the results to be expected was obtained by momentary overloading of a few smooth specimens before testing. These specimens were brought to temperature in a creep unit and weights were added until the desired amount of strain had been introduced. The excess weights were then quickly removed to give the starting stress for the test.

For S-816, Figure 26, prior straining of from 2.65 to 4 percent caused a markedly faster drop of stress level at early times. At the end of the first hour, residual stresses were some 5000 to 10,000 psi lower than they had been for specimens not previously overloaded. At longer times, the effects of prior overloading gradually became less noticeable and seemed to have disappeared by the end of about 100 hours.

Smaller initial plastic strains used in tests on the other two alloys, Figure 27, had little effect on relaxation rates for the conditions studied. However, what small changes did appear to result from prior straining were always in the direction of faster relaxation. Considering all the results at hand, it appears safe to conclude that plastic deformations up to a few percent of strain

should not interfere with relaxation of stresses in notched bars of the materials being considered, and would probably even promote relaxation in the case of S-816 at 1350°F.

Since relaxation is really a creep process in which parts of the specimen are constrained, any treatment which accelerates the relaxation process should similarly increase creep under steady stress. Creep data obtained during the first 5 to 10 minutes of various tests with S-816 have been assembled in Figure 28 for specimens with and without initial plastic straining by momentary overloading to 60,000 psi. Four prestrained specimens indicate a very definite increase in creep rates above those for other specimens at the same stresses but not overloaded prior to the test. Data at longer test times are fragmentary, but other pairs of tests at 40,000 and 36,200 psi showed the curves for prestrained specimens still to be drawing away from those without prestrain at the end of 2 hours. At that elapsed time the total creep for the four tests may be tabulated.

TABLE 1. COMPARATIVE TOTAL CREEP OF S-816 AT 1350°F DURING THE FIRST TWO HOURS WITH AND WITHOUT PRIOR PASTIC STRAIN

Spec. No.	Initial Plastic Strain and How Obtained	Initial Stress (psi)	Total Creep in 1st 2 Hrs (in./in.)
OS-S19	0.0048; momentary loading to 50,000 psi	40,000	0.00925
S-S20	0.00015	40,000	0.00725
OS-BS42	0.0293; momentary loading to 60,000 psi	36,200	0.00645
S-S22	0.00014	36,200	0.00340

While these tests may not constitute proof that S-816 should be expected to relax more rapidly after prior straining, the period over which an increased creep rate was obtained by initial overloading appears to be long enough to account for the period of very rapid initial relaxation observed for specimens deliberately prestrained, as well as for those which exceeded the proportional limit just by loading to the starting stress.

Creep to Rupture Under Single and Multiple Stress Levels

Among the basic data needed for the proposed analysis are creep and rupture properties for stresses expected in a notched bar. Such conventional creep curves, each run at a single stress level, are included on Figures 29 through 31. Table 2 lists times until rupture for these tests, along with elongation and reduction of area values. (See page 13.)

Rupture times have been plotted in Figure 32, including data of Carlson and Simmons mentioned previously (Ref. 9).

TABLE 2. STRESS - RUPTURE TIME DATA OBTAINED

Spec. No.	Stress (psi)	Rupture Time (hours)	Elongation[a] (percent)	Reduction of Area (percent)
		S-816 at 1350°F		
S-S10	65,000	1.2	37	39.5
S-S17	55,000	4.25	34	49.5
S-S13	45,000	25.2(±0.8)	52	58
S-S20	40,000	77.3(±2)	--	54
S-S22	36,200	142.8	43	53
S-S9	35,000	180.6	38	51.5
		Waspaloy at 1500°F		
S-W171	70,000	0.10	6.5	9.5
S-W157	60,000	0.50	7.5	10.5
S-W175	50,000	2.4	7.5	11.5
S-W163	40,000	10.15	7.5	9.5
S-W162	30,000	65.6	9	--
S-W174	23,000	292.1	9	12
S-W161	20,000	498.9	12	13
S-W173	17,000	1129.8	7	10
		Inconel X-550 at 1350°F		
S-X512	80,000	0.84	4.5	7.5
S-X504	70,000	4.6	5	6
S-X506	70,000	7.2	4.5	6.5
S-X511	60,000	36.7	2.5	5.5
S-X505	50,000	161.4	1	3
S-X509	35,000	1646.9	0.5	0.5

[a] Based on gauge length of 2.1 inches.

To be able to estimate the life expectancy of a fiber in a notched tensile bar in the manner proposed for this project, one must know not only the changes in stress which occur, but also what fraction of total life is expended by a given sojourn at each stress level. It would appear that expenditure of life should be related either to the length of time a stress has been acting at elevated temperature, or to the amount of creep which has taken place at the temperature and stress, or perhaps to some combination of time and stress.

Guarnieri and Yerkovich (Ref. 11) have suggested a method for handling periodic overstressing which amounts to adding for each stress the fraction

$$\frac{\text{actual creep elongation at the stress}}{\text{elongation to rupture at the stress}} .$$

In a second method, proposed by Robinson (Ref. 12) without supporting data, fractions added are equal to the ratio

$$\frac{\text{actual time at a given stress level}}{\text{rupture life at the stress in a conventional constant-load test}}$$

Both these methods for adding portions of life become the same for a material with uniform elongation at rupture over the range of stresses of concern, and for which the creep curves are all of the same type, i.e., with the same proportion of primary, secondary and tertiary creep from curve to curve. Of the three alloys studied, two had quite uniform elongation at rupture for the stress ranges investigated. (The variation of elongation for S-816 at 1350°F was from 39 to 58 percent; and for Waspaloy at 1500°F, from 6.5 to 12 percent.)

In contrast, the Inconel X-550 data showed a ten-fold spread in percent elongation. Results of multiple-stress tests for Inconel X-550 at 1350°F could indicate the relative applicability of strain fractions and of time fractions as a measure of life used up at any stress in a variable stress history. Eight tests were performed with Inconel X-550 in which the stress level was changed to a second value part way through the test. In Table 3 (see page 15), experimental results are compared with calculations based on rupture and elongation data (Figures 32 and 33) for single-stress rupture tests.

Some doubt exists as to the best curve at low stresses in Figure 33, but even allowing for the probable scatter from this cause, the calculations using strain as a measure of life expended seemed to err consistently on the side of predicting too long a life. On the other hand, the use of time fractions gave results which scattered nearly equally in both the high and low directions, with a maximum deviation observed of some 25 percent.

Data for Table 4 (see page 16) for S-816 and Waspaloy further support the addibility of time fractions of rupture life. These results do not, however, constitute an argument against the use of strain as a measure of life used up, as has been mentioned above.

A few representative multi-stress creep curves have been included on Figures 29, 30 and 31. These plots offer at least qualitative support for the finding that time fractions of rupture life are additive. The shape of the creep curve at a given stress, and after a given fraction of total life is gone, seems to be just about the same beyond that fraction of life, regardless of the stresses at which this fraction of life was consumed.

Considering all 15 multiple-stress rupture tests for the three materials, the maximum discrepancy of 26 percent for any single test is within the scatter expected for the usual single-stress test to rupture. Further, if the results for all 15 tests are averaged, addition of fractions of rupture lives checks the experiments quantitatively within one percent. It is concluded that for variable-stress tests of the materials under study the portion of life used up during a period of time at any particular stress is equal to the fraction

$$\frac{\text{Actual time at the given stress}}{\text{Rupture life for that stress}} .$$

TABLE 3. MULTIPLE-STRESS CREEP AND RUPTURE DATA FOR INCONEL X-550 AT 1350°F

Spec. No.	Stress Level (psi)	Time at Stress (hours)	[a]Creep at Stress (percent)	Actual Creep Elongation at Stress / Elongation to Rupture for Stress			Actual Time at Stress / Rupture Life for Stress		
MS-X513	50,000	48.6	0.06	0.06/1.0 =	0.06	) 1.61	48.6/122 =	0.40	) 0.73
	70,000	2.0	(6.4)	6.4/4.1 =	1.56		2.0/6.0 =	0.33	
MS-X514	70,000	2.0	0.37	0.37/4.1 =	0.09	) 1.69	2.0/6.0 =	0.33	) 1.05
	50,000	88.0	(1.6)	1.6/1.0 =	1.60		88.0/122 =	0.72	
MS-X517	44,300	200.5(±6)	0.17	0.17/0.88 =	0.19	) 1.49	200.5/280 =	0.72	) 0.83
	50,000	13.6	(1.3)	1.3/1.0 =	1.30		13.6/122 =	0.11	
MS-X518	53,680	40.8	0.15	0.15/1.55 =	0.10	) 1.16	40.8/75 =	0.54	) 1.22
	40,000	392.9	(0.85)	0.85/0.8 =	1.06		392.9/570 =	0.68	
MS-X519	47,410	166.0	0.16	0.16/0.95 =	0.17	) 0.61	166/185 =	0.90	) 1.26
	40,000	204.3	(0.35)	0.35/0.8 =	0.44		204.3/570 =	0.36	
MS-X520	41,190	321.4	0.10	0.1/0.82 =	0.12	) 2.02	321.4/420 =	0.77	) 1.16
	50,000	47.8	(1.9)	1.9/1.0 =	1.90		47.8/122 =	0.39	
MS-X521	56,820	20.9	0.25	0.25/2.0 =	0.12	) 1.06	20.9/50 =	0.42	) 0.89
	40,000	265.9	(0.75)	0.75/0.8 =	0.94		265.9/570 =	0.47	
MS-X522	38,060	404.6	0.07	0.07/0.75 =	0.09	) 1.09	404.6/670 =	0.60	) 1.06
	50,000	56.3	(1.0)	1.0/1.0 =	1.00		56.3/122 =	0.46	
					Avg:	1.34		Avg:	1.03

[a] Creep at second stress obtained from difference of elongation at rupture and creep which occurred at the initial stress.

TABLE 4. ADDITIONAL MULTIPLE-STRESS CREEP AND RUPTURE DATA

Spec. No.	Stress Level (psi)	Time at Stress (hours)	Creep at Stress[a] (percent)	Actual Time at Stress / Rupture Life at Stress	
			S-816 at 1350°F		
MS-S8	45,000	9.8	9.1	9.8/32 = 0.31	
	35,000	93.9	(37.4)	93.9/210 = 0.45	0.76
MS-S12	55,000	1.33	5.1	1.33/4.8 = 0.23	
	45,000	8.42	6.0	8.42/32 = 0.26	0.92
	35,000	80.05	(37)	80.05/210 = 0.36	
MS-S16	35,000	68.0	6.7	68/210 = 0.32	
	45,000	10.0	9.2	10.0/32 = 0.31	0.91
	35,000	56.2	(19.2)	56.2/210 = 0.27	
				Avg:	0.86
			Waspaloy at 1500°F		
MS-W158	60,000	0.177	0.8	0.177/0.46 = 0.38	
	30,000	34.0	(5.7)	34.0/215 = 0.76	1.14
MS-W164	40,000	3.4	0.9	3.4/8.0 = 0.42	
	30,000	22.4	1.2	22.4/45 = 0.50	1.23
	20,000	164.5	(3.4)	164.5/525 = 0.31	
MS-W165	20,000	107.3	0.2	107.3/525 = 0.21	
	40,000	3.4	4.0	3.4/8.0 = 0.42	0.97
	20,000	177.2	(4.3)	177.2/525 = 0.34	
MS-W166	40,000	5.1	2.15	5.1/8.0 = 0.64	
	20,000	287.3	(4.8)	287.3/525 = 0.55	1.19
MS-W168	20,000	165.0	0.3	165/525 = 0.315	
	40,000	2.5	3.4	2.5/8.0 = 0.315	0.98
	20,000	182.4	(3.7)	182.4/525 = 0.35	

TABLE 4. (Cont'd)

ADDITIONAL MULTIPLE-STRESS CREEP AND RUPTURE DATA

Spec. No.	Stress Level (psi)	Time at Stress (hours)	Creep at Stress[a] (percent)	Actual Time at Stress / Rupture Life at Stress	
MS-W169	10,000	215.0	0.025	215/33,500 = 0.01	0.99
	30,000	44.2	(11.0)	44.2/45 = 0.99	
MS-W172	18,000	263.2	0.4	263.2/1000 = 0.26	
	30,000	22.67	4.7	22.67/45 = 0.51	0.94
	18,000	172.9	(3.9)	172.9/1000 = 0.17	
				Avg:	1.06

a The creep for the final stress obtained from difference between elongation at rupture and creep which occurred at initial stress(es).

A Relationship Between Creep and Relaxation Properties

The quantitative nature of the results established in the previous section gives them far-reaching significance. If the law established for the limited range of variables investigated can be extended to higher and lower stress ranges, then it should be possible to predict relaxation characteristics directly from a family of creep curves. Relaxation properties could then be established at stresses above those feasible for study with the step-down method using known equipment.

Most attempts to correlate creep and relaxation have centered about so-called time-hardening and strain-hardening assumptions. In the former, creep rate is taken to be a function only of the variable stress and the total elapsed time, while the latter assumes creep rate to vary only with the stress and accumulated plastic strain. Figure 34 illustrates these two rules for a simplified case with four discrete levels in a step-down test with an exaggerated creep of one percent at each stress in turn.

The life-fraction rule substantiated above can also be applied to these hypothetical curves:

1. One percent strain at the stress S_1 takes two hours. This is 20 percent of the 10-hour rupture life assumed for S_1.

2. The new creep curve at the next stress S_2 begins at 20 percent of its rupture life (20 hours) or at a total time of four hours. One percent strain along this curve ends at 8.6 hours. Therefore, the life "used up" at S_2 is for (8.6 - 4) = 4.6 hours out of 20 hours rupture life, or about 23 percent. Now the total life which has expired is 0.20 + 0.23 = 0.43.

3. At stress S_3 the new creep segment begins at (0.43) (40 hours) = 17.2 hours. One percent of strain uses up 29 percent more life.

4. Continuing the procedure to the last stress S_4, after 4 percent total creep the specimen ends up at the 73-hour point on this curve, with approximately 7 hours of life remaining.

In comparison, the time-hardening and strain-hardening rules indicate that the 4 percent total strain corresponds respectively to about the 40- and 80-hour points on the S_4 curve.

The data for Inconel X-550 relaxing at 1350°F from an initial stress of 60,000 psi presented an unusual opportunity to check for a quantitative relationship since creep data were already on hand (see Figure 35) and since the relaxation had been slow enough that the early stages of the curve were known fairly accurately. The three suggested rules for predicting relaxation behavior from creep data have been applied to the stress levels used experimentally with relaxation specimen RS-X501, data for which were included on Figure 25. Details of the calculations are included in Appendix I. Pertinent results are summarized in Table 5 below.

TABLE 5. EXPERIMENTAL AND CALCULATED RELAXATION BEHAVIOR FOR INCONEL X-550 AT 1350°F FOR AN INITIAL STRESS OF 60,050 PSI

Residual Stress (psi)	Elapsed Time (hours)			
	Experimental	Strain-Hardening Rule	Life Fraction Rule	Time-Hardening Rule
60,050	0	0	0	0
56,820	0.37	0.4	0.4	0.4
53,680	1.53	1.4	1.2	1.2
50,550	5.5	5.8	5.2	5.0
47,410	15.1	16.6	16.0	15.5
44,300	35.0	33.1	33	31.5
41,190	66.0	48.1	52	57.8
38,060	115	73.6	114	123
34,910	198	106	164	224.5
31,790	321	181	294	498

As might be anticipated from the hypothetical case depicted in Figure 34, all three rules give substantially identical answers at high stresses where the curvature of the creep curves is slight. At longer times the three solutions spread widely, with the error for the time-hardening and strain-hardening assumptions tending in opposite directions from the experimental values. The life-fraction rule, in contrast, agrees remarkably well with the test data.

Perhaps a more convincing demonstration is offered by a pair of relaxation tests performed with Waspaloy. Each was started from an initial stress of 40,000 psi, but one was allowed to creep at that stress for 4.37 hours (0.0206 inches/inch creep strain) before the step-down relaxation process was begun. Creep curves for 40,000, 30,000, and 20,000 psi have already been reported. They are re-drawn to larger scale in Figure 36. An intermediate curve for 25,000 psi has also been added by the following expedient: Specimen RS-W176 had been allowed to creep to rupture at 25,000 psi after relaxation to that stress from 50,000 psi. The sum of fractions of life used up at the several stresses in the step-down relaxation test was about 10 percent, which is equivalent to 15 hours at 25,000 psi. If the proper strain at 15 hours were known, the creep curve for 25,000 psi could be continued from the point so located. The curve shown in Figure 30 was drawn through the 0.00114 inches/inch total strain observed for the cumulative relaxation steps, even though this is probably somewhat greater than the creep which would occur in 15 hours at a constant stress of 25,000 psi.

Appendix II gives details of the calculations involved to obtain the results of Table 6 (see page 20), which are also shown graphically on Figure 37. Agreement between experimental and calculated results is satisfactory whether or not extensive prior creep occurred before the relaxation run.

In a recent paper (Ref. 13), Roberts reported that the strain-hardening assumption yields accurate relaxation results from creep data for carbon steel and for S-816. In view of statements made previously, this is not in conflict with the present finding. For both these materials, variation in

TABLE 6. COMPARISON OF EXPERIMENTAL RELAXATION DATA FOR WASPALOY AT 40,000 PSI AND 1500°F WITH RELAXATION CURVES PREDICTED FROM CREEP DATA

Residual Stress (psi)	Elapsed Time to Relax to Indicated Residual Stress, hours: Relaxation after 4.37 hours Prior Creep at 40,000 psi (0.0206 in./in. Creep Strain)		No Prior Creep Before Relaxation Run	
	Experimental	Calculated	Experimental	Calculated
40,000	0	0	0	0
35,000	0.03	--	0.12	--
30,000	0.14	0.165	0.42	0.34
25,000	0.45	0.665	1.45	1.46
20,000	1.7	2.6	5.25	7.5
15,000	6.1	6.4	22.0	21.9

elongation for a large range of rupture stresses is probably too small to distinguish differences between results obtained using the strain-hardening and life-fraction rules.

Estimate of Portion of Life Consumed During Typical Relaxations

Despite the apparent validity of the life-fraction rule for runs with two or three discrete stress levels during a test, it seemed advisable to compare the remaining rupture life after a relaxation with that predicted by the rule. The first check tests were with S-816 at 1350°F, for which alloy and temperature the relaxation is so rapid that only a very small fraction of the total life should be used up during the relaxation period.

Specimen RS-S7 was first relaxed from 50,000 psi initial stress to 30,000 psi in 0.325 hours. The specimen lasted at the lower stress an additional 675.4 hours before it failed. This is even more than the normal rupture life of about 580 hours at 30,000 psi.

A second specimen (RRRS-S21) was subjected to three successive relaxations from 40,000, followed by creep until rupture at 30,000 psi, with the temperature maintained at 1350°F for the entire series. Two relaxation runs from 40,000 psi to a residual stress of 10,000 plus the third run down to 30,000 psi were estimated to have used up only about one percent of the available life. This was substantiated when the rupture test at 30,000 psi lasted an additional 669.1 hours. The mere fact that both these tests lasted somewhat longer than anticipated is not considered significant in that the value read from Figure 32 is very sensitive to slight changes in fitting the curve to the test points. The important conclusion seems to be that a brief period of high-stress relaxation has little detrimental effect on rupture life of S-816 at 1350°F, in agreement with prediction.

Six specimens of Inconel X-550 at 1350°F were similarly run to rupture after relaxation, again without intermediate cooling between parts of the test. In this case, however, the time required for relaxation is a significant part of the rupture life at the stresses involved. The relaxation period may be expected to consume a substantial portion of the available life of the material. Findings are tabulated for easy comparison. (See Table 7, page 22.)

Agreement between predicted and actual results is noteworthy when one considers that many of the rupture times used for these calculations involve extrapolations of a cycle or more on Figure 32.

A wide variety of alloys has been reported to show a greater resistance to relaxation when the same specimen is re-run after a first relaxation test. In the data of Robinson (Ref. 10), at least nine sets of data show such an apparent strengthening and only two pairs of tests give any hint of the reverse tendency. It is suggested that the "strengthening" often observed is only typical of relaxations occurring in the primary or decreasing-rate portion of the creep curves involved. A second relaxation may be slower, the same rate, or faster, depending on the character of the creep at the stresses and portion of life concerned. Two repeated-relaxation tests performed under this contract (Figure 38) support this contention. In these experiments, after a specimen had been subjected to one relaxation test, it was reloaded to the initial stress and allowed to relax again, all at the same test temperature.

TABLE 7. RUPTURE TESTS ON INCONEL X-550 AFTER PRIOR RELAXATION

Spec. No.	Stress-Time History	Summation of Fractions: Time at Given Stress / Rupture Life at Stress	
RRRS-X500	1) Load to 50,000 psi, relax to 25,300 psi (905 hrs)	0.97	
	2) Reload to 50,000 psi, relax to 28,390 psi (119.7 hrs)	0.10	
	3) Reload to 50,000 psi, relax to 37,670 psi (4.5 hrs)	0.14	
	4) Allow to creep to rupture at 37,670 psi (4.3 hrs)	0.13	
			1.10
RR-X501	1) Load to 60,050 psi, relax to 22,360 psi (851 hrs)	0.84	
	2) Reload to 60,050, relax to 38,060 psi (16.1 hrs) (Rupture at this stress.)	0.04	
			0.88
RS-X502	1) Relax 70,000 psi to 21,850 psi (758 hrs)	1.06	
	2) Reload to 70,000 psi and creep to rupture (0.22 hrs)	0.04	
			1.10
RS-X503	1) Relax 80,000 psi to 19,300 psi (857 hrs)	0.77	
	2) Reload to 50,000 psi and creep to rupture (1.6 hrs)	0.01	
			0.88
ORS-X507	1) Load to 89,760 psi and unload (0.0012 inches/inch total plastic strain)	0.04	
	2) Reload to 49,750 psi, relax to 24,210 psi (764 hrs)	0.80	
	3) Reload to 50,000 psi and creep to rupture (0.7 hrs)	0.01	
			0.85
RS-X508	1) Relax 70,000 psi to 35,000 psi (224 hrs)	0.85	
	2) Creep to rupture at 35,000 psi (344.5 hrs)	0.30	
			1.15
		Avg:	0.98

For the S-816, all three relaxations took place at the start of the several creep curves. Only at the very beginning of the first run was the rate apparently higher than for the other loadings. This may be a reflection of the much faster creep commonly observed when a specimen is first loaded or it may be the result of unavoidable experimental error in determining the very short times involved.

The situation for the Inconel X-550 experiments is in sharp contrast. The first relaxation alone accounted for most of the life of the specimen. In the repeated loadings, the material was in the third stage of creep. Consequently in each run the stress level dropped faster than it did the previous time. A long period of stress relaxation appears to harm Inconel X-550. This difference in behavior might be expected to be a key factor in the different notch-bar properties of the two alloys at this temperature.

We are now in position to distinguish between the effects of overloading S-816 before testing and similarly overloading Waspaloy or Inconel X-550. The latter two alloys have very little primary creep. The small change in

relaxation rate which appeared to result from prestraining of Waspaloy or Inconel X-550 was probably the simple result of using up a portion of the early life, during which time creep is slowest. On the other hand, the accelerated creep and relaxation of S-816 with prestrain was too large to have resulted from this latter cause.

Some Additional Checks on Addibility of Rupture Lives.

At the time it was desired to start tests on the effect of a period of relaxation on the life still remaining, some of the available Waspaloy specimens on hand had already been cooled and removed from the test units. This stress-time-temperature history differs considerably from that expected in the fibers of a notched bar, but rupture tests on such specimens were believed to be of probable value. Results obtained, including those for specimens with intermediate cooling, are tabulated below. (See Table 8, page 24).

From these data it appears that after a relaxation period of 150 hours or more, Waspaloy may retain only about half the rupture life normally expected for a given high stress level. In all but two cases the total life obtained was considerably below that expected. This seems to be contrary to findings in the multiple-stress tests described previously for the same alloy and temperature.

The chief difference between these tests and others performed in this investigation appears to be the extended relaxation times required to reach very low residual stresses. No completely satisfactory explanation has been found for the low rupture strengths obtained for these runs, but the possibility of metallurgical instability of Waspaloy over long periods at 1500°F warrants further consideration in the section on metallurgical factors.

TABLE 8. RUPTURE LIFE AT 1500°F OF WASPALOY SPECIMENS AFTER A PRIOR RELAXATION TEST

Spec. No.		Summation of Fractions: Time at Given Stress / Rupture Life at Stress	
RS-W151	1) Relax 40,000 to 4,820 psi at 1500°F (538 hrs) Unload and cool to room temperature	0.15	
	2) Reheat to 1500°F, load to 30,000 psi, creep to rupture (30.5 hrs)	0.68	
			0.83
SRS-W152	1) Creep 4.37 hrs at 40,000 psi, 1500°F	0.55	
	2) Relax to 4,850 psi (227.4 hrs), unload and cool to room temperature	0.06	
	3) Reheat to 1500°F, load to 35,000 psi, creep to rupture (7.2 hrs)	0.39	
			1.00
RS-W153	1) Relax 50,000 to 9,810 psi at 1500°F 9259 hrs) Unload and cool to room temperature	0.14	
	2) Reheat to 1500°F, load to 35,000 psi, creep to rupture (10.5 hrs)	0.57	
			0.71
ORS-W154	1) Heat to 1500°F. Turn off and cool to room temperature (poor temperature distributinn)		
	2) Reheat to 1500°F, in another unit, overload to 63,100 psi, unload to 40,000 psi	0.08	
	3) Relax to 10,000 psi (151.7 hrs)	0.09	
	4) Reload to 30,000 psi, creep to rupture (26.3 hrs)	0.59	
			0.76
RS-W156	1) Load to 70,000 psi at 1500°F (0.00140 inches/inch plastic strain	0.11	
	2) Relax to 3,800 psi (176 hrs)	0.17	
	3) Reload to 40,250 psi, creep to rupture (2.73 hrs)	0.36	
			0.64
ORS-W159	1) Overload to 81,600 psi at 1500°F. Unload to 40,500 psi (0.0149 inches/inch total strain)	0.83 (±0.2)	
	2) Relax to 6,400 psi (81 hrs)	0.04	
	3) Reload to 30,200 psi, creep to rupture (10.7 hrs)	0.24	
			1.11(±0.2)
		Avg:	0.84

SECTION III

COMPARISON OF NOTCHED-BAR RUPTURE BEHAVIOR WITH RELAXATION PROPERTIES

Conditions chosen for study in this investigation have since been shown (Ref. 9) to give two different types of notched behavior. Two of the alloys (S-816 at 1350°F and Waspaloy at 1500°F) exhibit notch strengthening over the entire range of stresses tested. The third material (Inconel X-550 at 1350°F) was strengthened slightly by a notch at high nominal stresses, but was increasingly weakened by a notch at lower applied stresses. It is sought to explain this observed difference in behavior in terms of tensile properties, creep and rupture data, and relaxation characteristics.

For the multi-axial stress condition around a notch, the effective stress ($\bar{S}$) causing yielding or controlling creep rates and rupture lives is tentatively assumed to be that indicated by the maximum shear-strain energy theory:

$$2\ (\bar{S}) = (S_a - S_h)^2 = (S_a - S_r)^2 + (S_h - S_r)^2\ , \qquad (1)$$

where S_a, S_h, and S_r are respectively the principal stresses in the axial, hoop, and radial directions of the specimen. Any occurrence which reduces the difference between any pair of these principal stresses will be reflected in a decreased effective stress, with consequent extension of life under the creep conditions present. Such a reduction in effective stress may result from the combined effects of plastic yielding near the notch root on loading the specimen and of creep relaxation during early stages of the test. The highest principal stress should relax faster than does the intermediate principal stress, while the smallest principal stress should decline the slowest of all. As the stress differences in Equation (1) become smaller through this mechanism, the effective stress must fall.

The cases of notch strengthening found above are hypothesized to represent materials such that initial plastic yielding plus relaxation of remaining concentrated stresses can reduce the effective stress below the nominal axial stress without using up very much of the alloy's total rupture life.

For Inconel X-550 at 1350°F, however, at the lower nominal stresses studied (about a half of the 0.2 percent offset yield strength) plastic yielding should be limited and relaxation is slow enough that a substantial amount of life should be consumed during the time period required for the effective stress to relax to a value below the nominal stress.

Outlines of a Proposed Method for a Quantitative Correlation of Notched-Bar Behavior and Smooth-Bar Properties

As mentioned in the Introduction, the basic premise being followed in this work is that a notch introduces nothing inherently new into properties of an alloy, but only changes the stress-strain histories of fibers in the notched bar. If one were to reproduce in a smooth bar the history experienced by any fiber of a notched bar, the life for each should be the same.

The problem resolves itself into two parts:

1. following the history of representative fibers in a notched bar, and
2. formulating general rules for total life of a fiber under various histories of stress and strain at the test temperature.

It is felt that the data presented in the previous sections of this report are sufficiently quantitative and extensive for treatment of any fiber loaded to stresses within the elastic range or up to plastic strains of the order of one percent. Initial stresses for portions of a notched bar within the elastic limit can be calculated from elasticity theory. For plastic portions, methods available are somewhat more approximate, but the errors made in estimating initial stresses and strains are not expected to seriously alter the final calculations.

For the purposes of discussion it will be assumed that the initial principal stresses and initial principal strains in the plane of a notch can be determined for fibers located at representative radii from the axis of a notched specimen. For each of such fibers the effective stress may be calculated and tabulated. Considering an appropriate time increment (eg., $\Delta T = 0.1$ hour), the decrease by relaxation of the effective stress in any fiber may be estimated. The resulting residual effective stress may be resolved again into components; addition of the axial components for all fibers gives the applied load which would be supported in the state of relaxed stress.

But the total axial thrust must not drop--it still must equal the applied tension. To bring the calculations back to actual conditions, the load "dropped" by relaxing fibers must be "picked up" again by the bar. This step may be imagined to be the same as an elastic addition of the same amount of load to a bar of the existing cross section and notch geometry. (As the process carries on, changes in notch root shape or in cross section at the base of the notch should occur by progressive creep. Suitable corrections are to be applied.)

When the principal stresses of all fibers have been rectified to give the proper total axial thrust, new values of the effective stress can be calculated. Each fiber has now withstood a known average stress for the length of time (ΔT) during which the above cycle occurred. Comparison with the conventional smooth-bar rupture life at this stress tells what portion of the total life of the fiber has been "used up."

The calculations are to be repeated for as many cycles as are necessary until the entire life of some fiber is calculated to have expired. The total elapsed time until this occurs may finally be compared with experimental rupture values for actual notched specimens.

If the effective stress fails to give a reasonable correlation, other combinations of the principal stresses may be tried. It might be noted that relaxation alone cannot account for the existence of notch strengthening if the maximum individual principal stress is the significant one determining rupture life.

Test Data Still Lacking for Proposed Quantitative Correlation Attempt

The critical fibers in a sharply notched bar are expected to be those immediately at the base of the notch or just below this surface. It is these particular fibers for which the initial strain is difficult to obtain when local stresses exceed the yield point. (The stress would be known from tensile data once the magnitude of strain is determined.)

The magnitude of strain on loading must also be known to determine whether additional relaxation and creep data are required, since effects on relaxation properties of more than about one percent strain have not been obtained in the present investigation.

Some studies of changes in notch shape when specimens are first loaded have been started by Carlson and Simmons as part of their work. Gross changes in shape appear to be small according to preliminary work, but microexamination of the material in the immediate vicinity of the notch root may show significantly-larger localized strains.

Until data about strains on loading are available, any attempt at a quantitative correlation is open to serious objection.

Qualitative Comparison of Notch Sensitivity and Relaxation Properties

The outstanding observation about the relaxation data obtained is the much slower rate for Inconel X-550 at 1350°F than for the other two alloys at the temperatures studied. It would be instructive to determine whether the order of difference in relaxation strengths observed is sufficient to account for differences in notch behavior reported by Carlson and Simmons.

In any sharp-notched specimen loaded to stresses of practical interest, localized yielding may be supposed to occur near the notch root. Of possible significance is the relative portions of life estimated to be expended by a fiber in a smooth bar while it relaxes from the 0.2 percent offset yield stress

to the stress at which rupture would occur in 1000 hours under steady load. Order-of-magnitude results are listed in Table 9 (see page 29) for the simplified case of three finite stress levels in a step-down process.

This tabulation (Table 9, page 29) suggests several points of interest:

1. Due to a low yield point and very rapid rates of relaxation of S-816 at 1350°F, any stress concentration is reduced so rapidly that very little life would be used up under any circumstances at the root of a notch. For these reasons, it would be almost impossible to have notch sensitivity in this alloy with the treatment used, as all data show. It was shown earlier that small amounts of yielding reduce resistance to relaxation for S-816. Thus the low yield strength contributes to reduction of stress concentrations even after all the load has been applied. The yielding and relaxation should help to reduce the effective stress below the nominal stress and thus prolong life.

2. Due to differences in yield strength and in rupture strength, Waspaloy would be required to relieve a much larger stress concentration than was S-816. Still its relaxation strength at 1500°F is low enough that only about 20 percent of Waspaloy's life would be used up in reducing the stress to the 1000-hour rupture strength. Further relaxation would then be expected to reduce the effective stress below the nominal and prolong life after about 10 hours. Thus it appears from these calculations that Waspaloy would have marginal notch sensitivity. The information at hand is not complete enough for accurate computations, as was previously discussed. The data of Carlson and Simmons (Figure 19) show that this alloy was slightly strengthened, an agreement considered very good for the rough estimates.

3. Inconel X-550 at 1350°F had much higher resistance to relaxation than was found for the other two cases. Combined with a larger differential between the yield and rupture strengths, this resulted in a substantial amount of life being used up at high stress levels. There seems little doubt that this alloy should be expected to show notch sensitivity at long periods, as has been found experimentally (Figure 18).

4. In all three cases the initial rate of relaxation was very rapid. According to the approximate calculations, even for Inconel X-550 at 1350°F, only seven percent of the life was used in reducing the stress of a smooth bar from the yield stress to 63,250 psi. This reduction occurred in 0.45 hours. Thus, for short-time, high-stress rupture tests, the effective stress would be reduced below the nominal rather rapidly with not too much life being used up. In such cases rupture times beyond those for smooth bars would be expected. This was found in tests.

Apparently at least two factors operate to show notch sensitivity at longer time periods, even though notch strengthening was found at short times. (1) The difference between the yield stress and the nominal stress is increased. (2) The material stays above the nominal stress for a longer time due to slow rate of relaxation at lower stresses.

5. The relation of elongation or reduction of area in the rupture test to notch sensitivity is not yet clear. Certain theories suggest themselves. One is that decreasing ductility with time to fracture is associated with a metallurgical change which increases resistance to relaxation. Another would be

TABLE 9. PORTION OF LIFE ESTIMATED TO BE EXPENDED BY A FIBER IN A SMOOTH BAR WHILE IT RELAXES FROM THE 0.2-PERCENT OFFSET YIELD STRESS TO THE 1000-HOUR RUPTURE STRESS FOR THE THREE ALLOYS STUDIED

Alloy and Temperature, °F		S-816 at 1350°F		Waspaloy at 1500°F		Inconel X-550 at 1350°F	
0.2% Offset Yield Stress, psi		47,500		76,500		89,500	
1000-hr Rupture Stress, psi		27,500		18,000		37,000	
Total Stress Decrement, psi		20,000		58,500		52,500	
Est. Time Required to Relax Each Third of Total Decrement, hours:	1st:	0.02		0.01		0.05	
	2nd:	0.03		0.07		0.4	
	3rd:	0.10		10		150	
Average Stress for Each Period, psi / Rupture Life at this Stress, hours:		Stress	Life	Stress	Life	Stress	Life
	1st:	44,150	38	66,750	0.2	80,750	1.05
	2nd:	37,450	140	47,250	3.3	63,250	20.5
	3rd:	30,800	490	27,750	75	45,750	230
Fraction of Life Expended for Each Period:	1st:	0.02/38 = 0.0005		0.01/0.2 = 0.05		0.05/1.05 = 0.05	
	2nd:	0.03/140 = 0.0002		0.07/3.3 = 0.02		0.4/20.5 = 0.02	
	3rd:	0.10/490 = 0.0002		10/75 = 0.13		150/230 = 0.65	
Total Fraction of Life Expended in Reaching 1000-hr Rupture Stress:		0.0009		0.20		0.72	

that the plastic deformation at the root of the notch alters structure in such a way as to reduce the amount the material can creep before fracture and therefore shortens rupture life. Possibly the pastic flow at the notch might alter a precipitation reaction in such a way as to reduce fracture strength.

6. It appears that relaxation rates for materials of the type considered and for the test temperatures employed are quite rapid in all cases. Thus, even for long-time tests, the damage caused by a notch occurs in a relatively short time. This may be due to the expending of a large proportion of the rupture life at high stresses and/or to alteration in properties due to the plastic flow involved in relieving stresses.

It follows, however, that as temperature is reduced and relaxation thereby slowed down, there should be a greater tendency for notch embrittlement.

SECTION IV

METALLURGICAL FACTORS INVESTIGATED

The compilation of notch rupture data in Section I indicated nearly complete lack of published information even as to which metallurgical factors are most important in affecting notch properties. The present program, therefore, studied a number of metallurgical variables in a somewhat cursory manner, rather than to delve into one or two aspects of the subject. Tests were limited to S-816 and Waspaloy as representing, respectively, a solid-solution type alloy an an age-hardenable alloy.

Foremost of interest was whether such metallurgical variations as are usually encountered in practice are apt to induce notch brittleness. It was assumed that standard alloys properly processed and heat-treated, and then tested at recommended service temperatures, should not be notch brittle. It is presumed that the alloys under consideration should not be evaluated for this purpose below 1500°F. Most of the tests were accordingly conducted at that temperature.

These metallurgical studies have been segregated into four categories for convenience of discussion:

1. Structural and property changes introduced during tests of alloys in the usual condition following conventional heat treatment.

2. Abnormal response to a standard heat treatment, reflecting effects of past history. The most obvious reason for such abnormal response is development of unusually-large grains during standard treatment. Other work at the University of Michigan had indicated that such excessive grain growth only results from critical deformations of the order of 0.5 to 2 percent. In practice such grain growth is encountered in forged blades for jet engines, due to difficulty in avoiding critical deformations.

Excessive growth of grains has been reported to be associated with low ductility which, in turn, has been shown to occur in many instances of notch sensitivity at elevated temperatures. It is also possible that some alloys may have varying notch characteristics, depending on hot-working conditions and response to heat treatment, even when the grain size is normal.

3. Deviations from recommended heat treatments, such as might occur by accident or through errors in pyrometric control. In actual practice, for different reasons, conditions of treatment occasionally deviate from those generally prescribed. Effects of such occurrences on notch properties should be useful information. Treatments above grain-coarsening temperatures would also be of value in showing the maximum effects to be expected.

4. Cold working or other extraneous treatments not usually included as part of the deliberate conditioning for the alloys under study. Finished parts may be subjected to additional operations which influence properties. One such operation would be cold straightening. In some cases, straightening is performed before aging or after partial aging. A few tests with different amounts of cold deformation performed at different points in the heat treatment were considered a necessary part of the program.

Surface-finish effects have been reported to have a profound effect on tests with sharp notches. This would appear to be a difficult problem to attack and has not been considered in the present broad program.

Structural Changes During Testing of Conventionally Heat-Treated Materials

In the study of notched-bar rupture properties, one must recognize that structural changes induced by the temperature, time at temperature, and stress conditions have to be integrated into the general explanation. Structural changes are important in the manner in which they change strength and ductility characteristics from those of the initial condition. Any particular changes in properties introduced by the special stress conditions associated with a notch would be especially important.

To date this subject has been investigated to only a limited extent by metallographic and hardness studies. The stress - rupture time curves showed changes in slope which are usually associated with structural changes during testing. The short rupture times for Waspaloy following prolonged relaxation (page 23) suggested property changes due to structural alterations during exposure to 1500°F. Likewise the very short first and second stages of creep found for Waspaloy specimens are indicative of structural instability.

During prolonged testing at 1350°F, S-816 underwent general precipitation and agglomeration normal for the alloy. (Compare Figures 14 and 39.) Photomicrographs (Figures 40 and 41) for the other two alloys after tests of long duration also show no drastic alteration of structure. Intergranular cracks were evident in all cases. Of probable significance was the absence of deep surface cracks for Inconel X-550, compared with the many such cracks at the surface of the other two alloys at fracture.

Thus far, hardness studies after testing have been confined to Waspaloy. Nine specimens involving testing times at 1500° F ranging from 3 to 1100 hours were subjected to hardness measurements with the results shown in Figure 42. Relaxation, multi-stress rupture and plain rupture tests have shown no consistent difference. Apparently the governing factor was time at temperature, with the hardness decreasing noticeably for times longer than 100 hours. Hardness ranged between Rockwell C values of 36 and 25. Sufficiently detailed studies have not been made for any conclusion other than that prolonged exposure of Waspaloy at 1500°F does result in a substantial reduction of hardness, presumably by overaging.

Rupture-Test Properties from Smooth and Notched Bars

After Abnormal Grain Growth

Grains as coarse as ASTM (-1) were obtained in both S-816 and Waspaloy by small reductions at room temperature prior to conventional heat treatments. Preliminary experiments showed that the extent of coarsening was erratic and not nearly so pronounced when specimens for cold reductions were taken from the stock just as received. Later specimens, including all those for which test data are reported, received the following initial treatments before critical cold rolling:

> S-816: 11.5 percent reduction at 1200°F + 2150°F, 1 hour, W.Q.
> Waspaloy: 45 percent reduction from 1950°F, A.C. + 1975°F, 1 hour, A.C.

For the S-816, one percent reduction appeared to give coarsest grains (from -1 to 2) in the gauge section of specimens made from the square bars. The surface material which was machined away tended to be somewhat finer (3 to 5). Waspaloy showed rather consistent abnormal coarsening for reductions between 1 and 1.5 percent prior to final heat treatment. A reduction of 1-1/4 percent appeared to be best, giving a uniform mixed grain size of (-1 to 1) and (2 to 6).

Typical photomicrographs for specimens with abnormally-large grains are shown in Figures 43 and 44.

Smooth specimens turned from bars of coarsened material had a reduced section about 1.6 inches long and a gauge diameter of 0.350 inches. Notched bars were prepared with a like minimum diameter (d). A root radius (r) of 0.004 inches was finished at the base of the 60° notch by a light cut with a hand-ground lathe tool.

Abnormal grain-growth response was accompanied by little or no change in rupture life for either smooth or notched bars of S-816 at 1350°F. Further, elongation of the smooth bars was not altered. (See Figure 45 and Table 10, page 34). In contrast, abnormal growth in Waspaloy resulted in a drop in ductility to about half the values for the material showing normal response, and the rupture life of Waspaloy for both smooth and notched bars appears to have been lengthened by treatments which results in large grains (Figure 45 and Table 11, page 35).

In the above comparisons, data of Carlson and Simmons (Ref. 9) were used as representative results for specimens with normal grain size. Carlson and Simmons used ground notches and grain size was not reported, but other bars from the same heats of materials with the same heat treatments had ASTM grain sizes from 4 to 7 for S-816; and from 3 to 6 for Waspaloy.

Reference to Table 10 (page 34) shows that S-816 specimen S-S52 came out with a very fine-grained structure. When this was tested as a smooth bar, results were essentially the same as for Specimen S-S50 with a (-1) to 1 grain size.

TABLE 10. EFFECT OF ABNORMAL GRAIN-GROWTH RESPONSE IN S-816 ON SMOOTH-BAR AND NOTCHED-BAR RUPTURE PROPERTIES AT 1500°F

Spec. No.	Stress (psi)	Life (hrs)	Elongation (%)	Reduction of Area (%)	Notch Specimen Design[a] D	d	r	r/d
Smooth Bars, Normal Grain Size - Data of Carlson and Simmons								
S-2	30,000	20.4	47.2	51.8				
S-4	25,000	63.8	44.8	58.5				
S-7	20,000	342.8	45.3	55.7				
S-11	17,000	1019.4	38.9	45.4				
Smooth Bars, Abnormal Grain Size (ASTM -1 to 2)								
S-S47	20,000	313.3+ (discontinued)						
S-S50	25,000	80.1	53	49				
S-S52[b]	25,000	88.3	45	49				
Notched Bars, Normal Grain Size - Data of Carlson and Simmons (Ref. 9)								
S-28	30,000	88.7	--	13.7	0.600	0.424	0.010	0.024
S-27	30,000	79.8	--	9.3	0.600	0.424	0.005	0.012
Notched Bars, Abnormal Grain Size (ASTM -1 to 2)								
N-S51	25,000	459.5			0.500	0.360	0.004	0.011
N-S53	25,000	428.4			0.500	0.360	0.004	0.011
N-S47	20,000	(1457) + (in progress)			0.500	0.360	0.004	0.011

[a] Dimensions of notches given in inches:
D = shank diameter of unnotched bar
d = diameter at notch root
r = root radius
Notch angle 60° in all cases.

[b] ASTM grain size 6 to 9.

In summary, no specimen either of Waspaloy or of S-816 appeared to be weakened by treatments which resulted in abnormal grain size upon later conventional heat treatment. This observation may not apply to lower temperatures or to other alloys, and these very limited data should not be construed as proof that there is no need for concern that abnormal grain size response may be accompanied by notch brittleness for the alloys under study. At most, they suggest that notch embrittlement may not be a function of grain size alone. Possibly some other unknown factor, accompanying the formation of large grains during heat treatment, is required for embrittlement.

TABLE 11. EFFECT OF ABNORMAL GRAIN-GROWTH RESPONSE IN WASPALOY ON SMOOTH-BAR AND NOTCHED-BAR RUPTURE PROPERTIES AT 1500°F

Spec. No.	Stress (psi)	Life (hrs)	Elongation (%)	Reduction of Area (%)	Notch Specimen Design[a] D	d	r	r/d
Smooth Bars, Abnormal Grain Size (ASTM -1 to +1, 2 to 4)								
S-W127	25,000	509.1	5.5	6.5				
S-W128	35,000	61.7	6.5	5				
S-W129	45,000	6.7	3	3				
S-W140	35,000	47.8	5	7.5				
S-W141	25,000	528.5	7.5	6.5				
S-W142	35,000	54.5	5	7.5				
S-W143	35,000	52.8	4.5	5.5				
S-W146	35,000	48.2	2	4				
Notched Bars, Normal Grain Size (ASTM 3 to 6)								
W-10[b]	35,000	72.5	--	2.3	0.600	0.424	0.040	0.094
W-20[b]	25,000	484.3	--	1.5	0.600	0.424	0.100	0.236
Notched Bars, Abnormal Grain Size (ASTM -1 to +1, 2 to 4)								
N-W131	40,000	150.3	--	--	0.350	0.265	0.003	0.011
N-W132	30,000	599.2	--	--	0.350	0.265	0.003	0.011
N-W135	25,000	1214.8	--	--	0.350	0.265	0.003	0.011

[a] Dimensions of notches given in inches:
D = shank diameter of unnotched bar
d = diameter at notch root
r = root radius
Notch angle 60° in all cases.

[b] Data of Carlson and Simmons, Ref. 9.

Effects on Notch Behavior of Deviations from Recommended Heat Treatments

S-816

Four smooth-bar tests were run on specimens of S-816 solution treated at 2325°F instead of 2150°F. Two of these specimens were given the customary 12-hour age at 1400°F, air cooled. The others were tested with no aging other than that incidental to heating the specimen for testing. Test results indicate a slight drop in ductility following the higher solution temperature.

This effect was augmented by omitting the age before testing. Variation in grain size with solution temperature was slight. Rupture data have been plotted in Figure 46.

High solution temperature seemed to have little significant effect on rupture life at 1350° or 1500°F. Elongation at rupture of smooth bars (Table 12 below) was sufficiently high that notch sensitivity is not to be expected at either temperature, although this has not been confirmed experimentally.

TABLE 12. SMOOTH-BAR RUPTURE TESTS WITH S-816 SOLUTION-TREATED AT 2325°F, 1 HOUR, WATER QUENCH

Spec. No.	Aging Time at 1400°F (hours)	Test Temp (°F)	Stress (psi)	Life (hrs)	Elongation (%)	Reduction of Area (%)	ASTM Grain Size
S-30	12	1350	35,000	317.5	40.0	38.5	(3), 4
S-31	12	1500	25,000	97.7	33.5	36	to 7
S-33	0	1350	35,000	308.5	24[a]	20.5	(3), 4
S-34	0	1500	25,000	71.3	20	18	to 6

[a] Broke in gauge mark.

Waspaloy

Effect of using a solution temperature different from the recommended value (1975°F) was investigated in somewhat greater detail for Waspaloy than for S-816. On the other hand, variation in aging practice was not studied at all. Data obtained at Pratt and Whitney Aircraft and reported by Simmons and Cross (Ref. 14) were considered satisfactory as an indication of the general effects of deviations from usual aging procedures.

A first consideration was the magnitude of effects introduced by temperature errors of moderate amount during heat treatment. Later tests aimed at determining how high a temperature is required before excessive grain coarsening sets in and to investigate whether conditions leading to such coarsening adversely alter notch strength. Rupture properties were determined for a limited number of notched specimens, as well as for smooth bars (Table 13, page 37).

There was no evidence of any large change in rupture properties at 1500°F for either smooth or notched bars as solution temperature was altered. (See Figure 47.) Notched-bar strength tended to decrease with increasing temperature. However, there was a substantially longer rupture time for notched specimens than for smooth at any stress studied. Ductility dropped somewhat, particularly for the longer-time tests of the coarse-grained material solution treated at 2150°F (Figure 48), suggesting that notch weakening might be encountered for longer times than those tested. It should be recognized that the indication of absence of notch weakening is contrary to reported experience for high solution temperatures. Apparently some additional factor must have been present to reduce notch sensitivity.

TABLE 13. VARIATION IN RUPTURE PROPERTIES WITH SOLUTION TEMPERATURE FOR SMOOTH AND NOTCHED BARS OF WASPALOY

(All tests at 1500°F. All specimens aged 1550°F, 4 hours, Air Cool + 1400°F, 16 hours, Air Cool.)

Spec. No.	Solution Temp (°F)	Grain Size (ASTM No.)	Stress (psi)	Rupture Life (hours)	Elong-ation (%)	Reduction of Area (%)
		Smooth Bars				
S-W177	1925		30,000	86.4	9.5	12
S-W163	1975	3 to 6	40,000	10.15	7.5	9.5
W-6[a]	1975	3 to 6	35,000	17.2	10.5	15.7
S-W162	1975	3 to 6	30,000	65.6	9	--
S-W186	2035	2 to 3, 3 to 6	30,000	65.8	8	8
S-W189	2075	(2), 3 to 5	35,000	25.9	6.5	7.5
S-W191	2150	(0), 1 to 3, (4 to 5)	35,000	41.4	5	6.5
S-W105-1[b]	2150	(-2), -1 to 0, 2 to 3	30,000	82.4	2.5	4
		Notched Bars[c]				
N-W178	1925		35,000	122.7		
W-10[a]	1975		35,000	72.5		
W-20[a]	1975		25,000	484.3		
N-W187	2035		35,000	101.1		
N-W190	2075		35,000	47.5		
N-W192	2150		35,000	66.5		

[a] Data of Carlson and Simmons (Ref. 9).

[b] Had been reduced 45 percent in 3 rollings from 1950°F, A.C. + 1975°F, 4 hrs, A.C. before solution treatment at 2150°F.

[c] Notch geometry of specimens was as follows, all dimensions in inches:

	Univ. of Mich. Data	Data of Carlson and Simmons
Shank Diameter (D)	0.500	0.600
Notch Root Diameter (d)	0.375	0.424
Root Radius (r)	0.004	W-10: 0.040 W-20: 0.100
Notch Angle	60°	60°

Effect of Extraneous Treatments on Notch Behavior

S-816

Changes in rupture properties were noted for specimens cold rolled after solution treatment either at the usual 2150°F or else at 2325°F. Treatments applied and results obtained are listed in Table 14 (see page 39) and plotted in Figure 49.

Cold rolling either at room temperature or at 1200°F raised smooth-bar rupture strengths some 10,000 to 20,000 psi above the usual values for the range of stress and temperature investigated. The data are not complete enough to show with certainty any large effect of aging at 1400°F between rolling and testing. It might have been anticipated that cold working should lower elongation at rupture, but some of the ductilities observed are even lower than expected. Of particular interest are the consistently-lower ductilities found at 1500°F test temperature than for tests conducted at 1350°F. Elongation at rupture was only one percent for two specimens and was five percent or less for most of the 1500°F tests.

Sufficient notched-bar data have not been obtained to permit generalizations. But for a test temperature of 1350°F specimens cold rolled after conventional solution treatment at 2150°F had low enough elongation in the rupture test that the material may be notch brittle. Elongation appeared to be even lower in tests at 1500°F. The very low elongation (1 percent) found at rupture for specimens with 2325°F solution temperature followed by 13.5 percent reduction at 1200°F indicates most probable notch weakening, in view of literature data surveyed in Part I of this report.

Cold rolled S-816 offers an opportunity to ascertain whether ductility has fundamental significance in determining notch behavior or whether relaxation properties are more truly significant. In any event, more extensive testing appears warranted.

A typical photomicrograph of the cold-rolled S-816 material used in the above tests is shown in Figure 50.

Waspaloy

Cold work was introduced into Waspaloy before any aging had taken place, after partial aging, or at the completion of aging. (Data listed also include some specimens reduced more than the critical amount in preliminary attempts to find conditions for abnormal grain growth. For these bars, all cold work was done before the solution treatment.)

Experimental results are assembled in Table 15 (see page 40) and Figures 51 and 52. A representative photomicrograph of a Waspaloy specimen after receiving an extraneous cold reduction during processing is shown in Figure 53.

TABLE 14. RUPTURE-TEST RESULTS FOR S-816 AFTER EXTRANEOUS TREATMENTS

Type Spec.	Test Temp (°F)	Stress (psi)	Rupture Life (hours)	Elongation (%)	Reduction of Area (%)
2150°F, 1 hr, W.Q. + 10% Reduct. at 75°F + 1400°F, 12 hrs, A.C. (ASTM Grain Size 5 to 8)					
Smooth	1350	45,000	105.2	35.5	50
Smooth	1500	25,000	318.5	5	6
2150°F, 1 hr, W.Q. + 10% Reduct. at 75°F (ASTM Grain Size 5 to 8)					
Smooth	1350	45,000	75.4	10.5	13.5
Smooth	1500	25,000	332.2	4.5	10
2325°F, 1 hr, W.Q. + 13.5% Reduct. at 1200°F, A.C. + 1400°F, 12 hrs, A.C. (ASTM Grain Size 4 to 6)					
Smooth	1350	45,000	476	3.5	13.5
Smooth	1350	35,000	(2478)+	In progress	
Smooth	1500	40,000	20.9	1	4.5
Smooth	1500	30,000	204.1	1	5
2325°F, 1 hr, W.Q. + 13.5% Reduct. at 1200°F, A.C. (ASTM Grain Size 4 to 6)					
Smooth	1350	35,000	1188.9	5	17
Smooth	1500	20,000	(3284)+	Discontinued	
Notched[a]	1500	35,000	8.05		
Notched[a]	1500	30,000	8.4		
2325°F, 1 hr, W.Q. + 5% Reduct. at 75°F (ASTM Grain Size 4 to 6)					
Smooth	1500	25,000	345.8	12	16.5
Notched[a]	1500	35,000	1[illegible].6		

[a] Notch geometry: Shank Diameter (d) = 0.500 inches
Diameter (d) at notch base = 0.375 inches
Root Radius (r) = 0.001 inches
Notch Angle = 60°

TABLE 15. RUPTURE-TEST DATA AT 1500°F FOR WASPALOY WITH EXTRANEOUS TREATMENTS

Spec. Type	Stress (psi)	Rupture Life (hours)	Elongation (%)	Reduction of Area (%)
1975°F, 4 hrs, AC + 5% Red. at 75°F + 1550°F, 4 hrs, AC + 1400°F, 16 hrs, AC				
Smooth	30,000	3.2 (± 3)	<1	2
Notched[a]	35,000	4.5	--	--
1975°F, 4 hrs, AC + 1550°F, 4 hrs, AC + 5% Red. at 75°F + 1400°F, 16 hrs, AC				
Smooth	25,000	38.0	1-1/4	2.5
Notched[a]	30,000	33.1 (± 2.5)	--	--
1975°F, 4 hrs, AC + 1550°F, 4 hrs, AC + 1400°F, 16 hrs, AC + 5% Red. at 75°F				
Smooth	25,000	95.3	2	2
Notched[a]	25,000	159.5 (± 1)	--	--
1975°F, 4 hrs, AC + 15% Red. at 75°F + 1550°F, 4 hrs, AC + 1400°F, 16 hrs, AC				
Smooth	30,000	5.8	1	1
Notched[a]	20,000	53.0	--	--
1975°F, 4 hrs, AC + 1550°F, 4 hrs, AC + 15% Red. at 75°F + 1400°F, 16 hrs, AC				
Smooth	35,000	2.6	1	1.5
Notched[a]	25,000	26.5	--	--
1975°F, 4 hrs, AC + 1550°F, 4 hrs, AC + 1400°F, 16 hrs, AC + 15% Red. at 75°F				
Smooth	40,000	3.7	<1	2.5
Notched[a]	25,000	46.4	--	--
10% Red. at 75°F + 1975°F, 4 hrs, AC + 1550°F, 4 hrs, AC + 1400°F, 16 hrs, AC				
Smooth	30,000	88.9	8.5	8
Smooth	25,000	260.5	7.5	9
1.5% Red. at 75°F + 1975°F, 4 hrs, AC + 1550°F, 4 hrs, AC + 1400°F, 16 hrs, AC				
Smooth	30,000	132.6	5	7
Smooth	25,000	479.1	7	7.5
Smooth	20,000	1203.3	3.5	7

[a] Notch Geometry: Shank diameter (d) = 0.500 inches; Diameter (d) at notch base = 0.375 inches; Root radius (r) = 0.004 inches; Notch angle = 60°

The tests, all conducted at 1500°F, show that:

1. Cold reductions of 5 percent and 15 percent after solution treatment lowered smooth-bar strength below that for conventional heat treatment. Very low ductility values also resulted after such cold-working treatments.

2. Though data on notched specimens of material cold worked after solution treatment are not sufficiently complete to be certain of effects, it may be noted that:

(a) Material reduced 5 percent had notch strengths above that of similarly-treated smooth specimens, in spite of very low elongation and reduction of area for the smooth bar tests.

(b) Material reduced 15 percent had notch strengths which appear to be marginal with respect to the corresponding smooth bars. Again there is little evidence of pronounced notch sensitivity in spite of very low ductility in the smooth bar tests (elongation of about 1 percent).

3. Cold work after aging was somewhat less damaging than before aging. Cold work after partial aging was intermediate in reducing strength.

4. Reductions prior to solution treatment of 1.5 and 10 percent raised smooth bar rupture strengths without much effect on ductility.

Summary of Metallurgical-Variable Studies

The metallurgical variables investigated to date have not caused definite notch sensitivity to occur in S-816 at 1350° or 1500°F, or in Waspaloy at 1500°F. In most cases, however, the results for notched specimens are very incomplete.

S-816 may be notch sensitive in respect to smooth bars given the same treatments when it is solution treated (particularly at a higher temperature than usual) and then cold worked. It is questionable, however, whether such materials would be notch sensitive in comparison to conventionally-treated smooth bars, due to the increase in strength from the cold work.

Waspaloy cold worked 5 and 15 percent after solution treatment showed a substantial loss in rupture strength and very low elongation and reduction of area in the rupture tests. Notched bars given these same treatments, however, presented little or no evidence of being notch sensitive in a few survey tests.

Critical reductions to develop grain sizes as large as ASTM (-1) upon normal solution treatment did not indicate notch embrittlement. Neither did raising the solution temperature of Waspaloy to as high as 2150°F.

All results to date suggest that it would be difficult to make either S-816 at 1350° or 1500°F, or Waspaloy at 1500°F notch sensitive. This result was not unexpected for S-816. The lack of notch sensitivity for Waspaloy

was somewhat surprising in view of reported tendency for the alloy to become notch sensitive without careful control of heat treatment. Either the particular heat of alloy used is unusually resistant to notch embrittlement, or 1500°F is too high for the alloy to become notch sensitive except under most unusual conditions. Notch sensitivity was shown at 1200° and 1350°F for the same Waspaloy material after normal treatment.

SECTION V

FUTURE WORK

The work planned for the future involves the following objectives:

1. Calculation of the influence of relaxation by creep on the total life of notched specimens. To do this, it will be necessary to measure the change in shape of notched specimens on loading and during the early stages of the creep-rupture tests. When these localized strains are less than one percent, calculations as outlined in Section III can be used. If initial strains are larger, tests will be necessary to establish the influence of the larger strains on relaxation behavior.

2. Obtain relaxation data for a material which has marginal notch sensitivity. Waspaloy at 1350°F has been selected for this purpose. Originally it was thought that Waspaloy at 1500°F would meet this requirement. The work of Carlson and Simmons now shows notch strengthening at 1500°F, with the marginal properties desired at 1350°F. Relaxation properties will be established at 1350°F as in the cases described in Section II.

3. Introduce variable notch properties for the same alloys with the same heat treatment. It is often reported that notch properties of the alloys in the program vary widely from heat to heat. It would be desirable to include a heat of one of the alloys which is more sensitive to notched and one less sensitive than the one already studied. Attempts will be made to procure such heats of Waspaloy, probably using vacuum melted material for the notch insensitive material.

4. Check the generality of the relaxation concept for explaining notch sensitivity in creep-rupture tests. This is being done in part by extending the work on metallurgical variables.

BIBLIOGRAPHY

1. Neuber, H. Theory of Notch Stresses. (Translated by T. A. Raven for the David Taylor Model Basin, United States Navy.) Edwards Brothers, Ann Arbor, Mich., 1946.

2. Badger, W. L. Progress Report for NACA Subcommittee on Heat-Resisting Materials. General Electric Company, 10 December 1951.

3. Data Sheets, dated 12 March 1951 and 21 February 1952. Allegheny Ludlum Steel Corporation. Supplied by Dr. G. Mohling.

4. Eiselstein, H. L. Notched Rupture Bar Testing, May 22, 1952 (C-3). Supplied by Mr. C. A. Crawford, Development and Research Division, The International Nickel Co., Inc.

5. Brown, W. F., Jr.; Jones, M. H.; and Newman, D. P. Influence of Sharp Notches on the Stress-Rupture Characteristics of Several Heat-Resisting Alloys. American Society for Testing Materials, Special Technical Publication No. 128, "Symposium on Strength and Ductility of Metals at Elevated Temperatures....", June 1953, pp. 25-45.

6. Davis, E. A. and Manjoine, M. J. Effect of Notch Geometry on Rupture Strength at Elevated Temperatures. American Society for Testing Materials, Special Technical Publication No. 128, "Symposium on Strength and Ductility of Metals at Elevated Temperatures...", June 1953, pp. 67-87.

7. Memorandum on Smooth and Notched Bar Rupture Testing of Timken 16-25-6 Alloy Wheel Rim Forgings; Thomson Laboratory, General Electric Company, 22 October 1952.

8. Sachs, G. and Brown, W. F., Jr. A Survey of Embrittlement and Notch Sensitivity of Heat-Resisting Steels. American Society for Testing Materials, Special Technical Publication No. 128, "Symposium on Strength and Ductility of Metals at Elevated Temperatures...", June 1953, pp. 6-20.

9. Carlson, R. L. and Simmons, W. F. Fifth Quarterly Progress Report on an Investigation on Notch Sensitivity of Heat-Resistant Alloys at Elevated Temperatures. Battelle Memorial Institute, 15 October 1953.

10. Robinson, E. L. High Temperature Bolting Materials. Proceedings of the American Society for Testing Materials. Vol. 48. (1948). pp. 214-238.

11. Guarnieri, G. J. and Yerkovich, L. A. The Influence of Periodic Overstressing on the Creep Properties of Several Heat Resistant Alloys. Proceedings of the American Society for Testing Materials. Vol. 52 (1952). pp. 934-950.

12. Robinson, E. L. Effect of Temperature Variation on the Long-Time Rupture Strength of Steels. Transactions of the American Society of Mechanical Engineers, Vol. 74, No. 5. July 1952. pp. 777-781.

13. Roberts, I. Prediction of Relaxation of Metals from Creep Data. Proceedings of the American Society for Testing Materials. Vol. 51 (1951). pp. 811-825.

14. Simmons, W. and Cross, H. Second Quarterly Progress Report on Investigation on Notch Sensitivity of Heat-Resistant Alloys at Elevated Temperatures. Battelle Memorial Institute, 8 December 1952.

APPENDIX I

CALCULATIONS FOR PREDICTION OF RELAXATION PROPERTIES OF INCONEL X-550 FROM CREEP CURVES FOR THE SAME TEMPERATURE

In the experimental step-down relaxation test at 1350°F with Inconel X-550 specimen RR-X 501, the strain corresponding to the 3140 psi stress decrement upon removal of each weight was 0.00013 inches per inch. Creep strain of approximately this amount was allowed to occur at each stress in turn from an initial value of 60,050 psi to a final value of 34,910 psi.

Below, the times required for these periods of creep strain at each of the several stresses used have been read from the creep curves of Figure 35 on the basis of three rules which have been proposed to correlate creep and relaxation properties.

A. Calculations Using the Time-Hardening Rule.

According to the time-hardening rule, when the stress is lowered from one level to another in a step-down test the new segment on the second creep curve begins at the time coordinate where that for the first stress terminated. (See Figure 34). Applying this rule to the creep curves of Figure 35, results may be tabulated as follows:

Creep Stress (psi)	Time at Start of Creep at Given Stress (hrs)	Time for Creep of 0.00013 in/in at Given Stress (hrs)	Cumulative Time by End of Each Creep Period (hrs)
60,050	0	[a] 0.4	0.4
56,820	0.4	1.2 - 0.4 = 0.8	1.2
53,680	1.2	5.0 - 1.2 = 3.8	5.0
50,530	5.0	15.5 - 5.0 = 10.5	15.5
47,410	15.5	31.5 - 15.5 = 16.0	31.5
44,300	31.5	57.8 - 31.5 = 26.3	57.8
41,190	57.8	123 - 57.8 = 63.2	123
38,060	123	224.5 - 123 = 101.5	224.5
34,910	224.5	498 - 224.5 = 273.5	498
(31,790)	498		

[a] Estimated from creep curve for 60,000 psi, Figure 31.

B. Calculations Using Strain-Hardening Rule.

By the strain-hardening rule, each successive segment of the creep curves involved in a step-down relaxation test is assumed to begin at the same strain as that where the segment terminated for the previous stress level. (See Figure 34.) This rule gives the following results:

Creep Stress (psi)	Cumulative Strain at Start of Given Creep Period (in/in)	Time for Creep of 0.00013 in/in at Given Stress (hrs)	Cumulative Time by End of Given Creep Period (hrs)
60,050	0	(0.4)[a]	0.4
56,820	0.00013	(1.0)	1.4
53,680	0.00026	11.5 - 7.1 = 5.8	5.8
50,530	0.00039	42.0 - 31.2 = 10.8	16.6
47,410	0.00052	80.3 - 63.8 = 16.5	33.1
44,300	0.00065	125 - 110 = 15	48.1
41,190	0.00078	317.5 - 292 = 25.5	73.6
38,060	0.00091	(32)[b]	106
34,910	0.00104	1255 - 1180 = 75[c]	181
(31,790)	0.00117		

[a] Estimated from creep curve for 60,000 psi, Figure 31.

[b] Creep data for 38,060 psi did not extend to the times required. Extrapolation indicated the required time increment at this stress to be about 27 hours for 0.00013 inches per inch creep. On the other hand, the log mean of the times which would be required for 0.00013 inches per inch creep for 41,180 psi and for 35,000 psi, each starting at the initial creep strain of 0.00091 inches per inch, was 37 hours. The mean value of 32 hours has been used.

[c] Read from a continuation of the curve not shown in Figure 35.

C. Calculations Assuming Additivity of Life Fractions.

The data tabulated on the next page were obtained from the creep curves of Figure 35 in the manner illustrated for the life-fraction rule in the hypothetical curves of Figure 34.

CALCULATION OF RELAXATION FOR INCONEL X-550 AT 1350°F

ASSUMING ADDITIVITY OF LIFE FRACTIONS

Creep Stress (psi)	Rupture Life at Given Stress (hrs)	[a]Time Coordinate at Start of Given Creep Period (hrs)	Time for Creep of 0.00013 in/in at Given Stress (hrs)	Fraction of Life Expended at Given Stress	Cumulative Fraction of Life Expended	Cumulative Time of Creep Periods (hrs)
60,050	35	0	0.4[b]	0.4/35 = 0.0114	0.0114	0.4
56,820	50	(0.0114)(50) = 0.57	1.37 - 0.57 = 0.8	0.8/50 = 0.016	0.0274	1.2
53,680	75	(0.0274)(75) = 2.1	6.1 - 2.1 = 4.0	4/74 = 0.053	0.080	5.2
50,530	115	(0.080)(115) = 9.2	20.0 - 9.2 = 10.8	10.8/115 = 0.094	0.174	16.0
47,410	185	(0.174)(185) = 32.2	49.2 - 32.3 = 17	17/185 = 0.095	0.27	33
44,300	280	(0.27)(280) = 75.5	94.5 - 75.5 = 19.0	19/280 = 0.063	0.34	52
41,190	420	(0.34)(420) = 143	195 - 143 = 62	62/420 = 0.148	0.49	114
38,060	670	(0.49)(670) = 328	378 - 328 = 50	50/670 = 0.08	0.57	164
34,910	1450	(0.57)(1450) = 830	(130)[c]	130/1450 = 0.09	0.66	294
(31,790)						

[a] Calculated as the product of the rupture life at the given stress in the conventional constant-stress rupture test times the total fraction of life already expended at previous stress levels in the step-down test.

[b] Estimated from creep curve for 60,000 psi, Figure 31.

[c] Estimated using slope of 35,000 psi creep curve at 300 hours.

APPENDIX II

CALCULATION OF RELAXATION CURVES FROM 40,000 TO 15,000 PSI FOR WASPALOY AT 1500°F WITH AND WITHOUT PRIOR CREEP AT THE INITIAL STRESS

Suitable data for early stages of creep of Waspaloy are given in Figure 36. The life-fraction rule (see Figure 34) will be applied to a step-down test with the first stress decrement taken as 10,000 psi followed by three stress reductions of 5000 psi each.

The following test points not shown for the 40,000 and 30,000 psi curves of Figure 36 are available for calculations at early times.

Specimen S-W163, 40,000 psi stress

Time (hrs)	Creep Strain (in/in)
0.1	0.00032
0.25	0.00053
1.1	0.00166
2.5	0.00410

Specimen S-W162, 30,000 psi stress

Time (hrs)	Creep Strain (in/in)
0.2	0.000171
1.0	0.000342

The experimentally-observed modulus of elasticity for Waspaloy at 1500°F was 21×10^6 psi/in/in. Therefore the creep strain required to give a 5000 psi reduction in stress levels is

$$\frac{5000 \text{ psi}}{21,000,000 \text{ psi/in/in}} = 0.00024 \text{ in/in}$$

This value of required amount of creep for each step (0.00048 in/in for the first step of 10,000 psi) has been used in the calculations tabulated on the following page.

In this tabulation, the time coordinate at the start of any given creep period was taken as the product of the cumulative fraction of life previously expended times the rupture life for the material in a conventional constant-load test at the current stress level.

The time for the required creep at the given stress level was calculated as the product of the reciprocal slope (hr/in/in) estimated from the creep curve at the starting time concerned times the required amount of creep (in/in) at the stress for the desired stress relaxation.

TABLE 16. CALCULATION OF RELAXATION PROPERTIES FROM CREEP DATA FOR WASPALOY AT 1500°F

Creep Stress (psi)	Rupture Life at Given Stress (hrs)	Time Coordinate at Start of Given Creep Period (hrs)	Time at Stress for Required Creep (hrs)	Fraction of Life Expended at Stress	Cumulative Fraction of Life Expended	Cumulative Relaxation Time at Start of Each Creep Period (hrs)
A. Relaxation with No Prior Creep at the Initial Stress						
40,000	11.7	0.0	(0.15/0.00021)*(0.00048)** = 0.34	0.34/11.7=0.029	0.029	0
30,000	55	(0.029)(55)=1.6	(0.08/0.00017)(0.00024) = 1.12	1.12/55=0.020	0.049	0.34
25,000	150	(0.049)(150)=7.5	(10/0.0004)(0.00024)=6	6/150-0.04	0.09	1.46
20,000 (15,000)	480	(0.09)(480)=43	(30/0.0005)(0.00024)=14.4	14.4/480=0.03	0.12	7.5 (21.9)
B. Relaxation after 4.37 hours Creep at 40,000 psi						
40,000	11.7	4.37	(1/0.0029)*(0.00048)** = 0.165	(4.37+0.165)/11.7 = 0.392	0.392	0
30,000	55	(0.392)(55)=21.5	(5/0.0024)(0.00024)=0.5	0.5/55=0.009	0.40	0.165
25,000	150	(0.40)(150)=60	(12/0.0015)(0.00024)=1.92	1.92/60=0.03	0.43	0.665
20,000 (15,000)	480	(0.43)(480)=206	(8/0.0005)(0.00024)=3.8	3.8/480=0.008	0.44	2.6 (6.4)

* Reciprocal slope of creep curve, hr/in/in.
** Required creep strain, in/in.

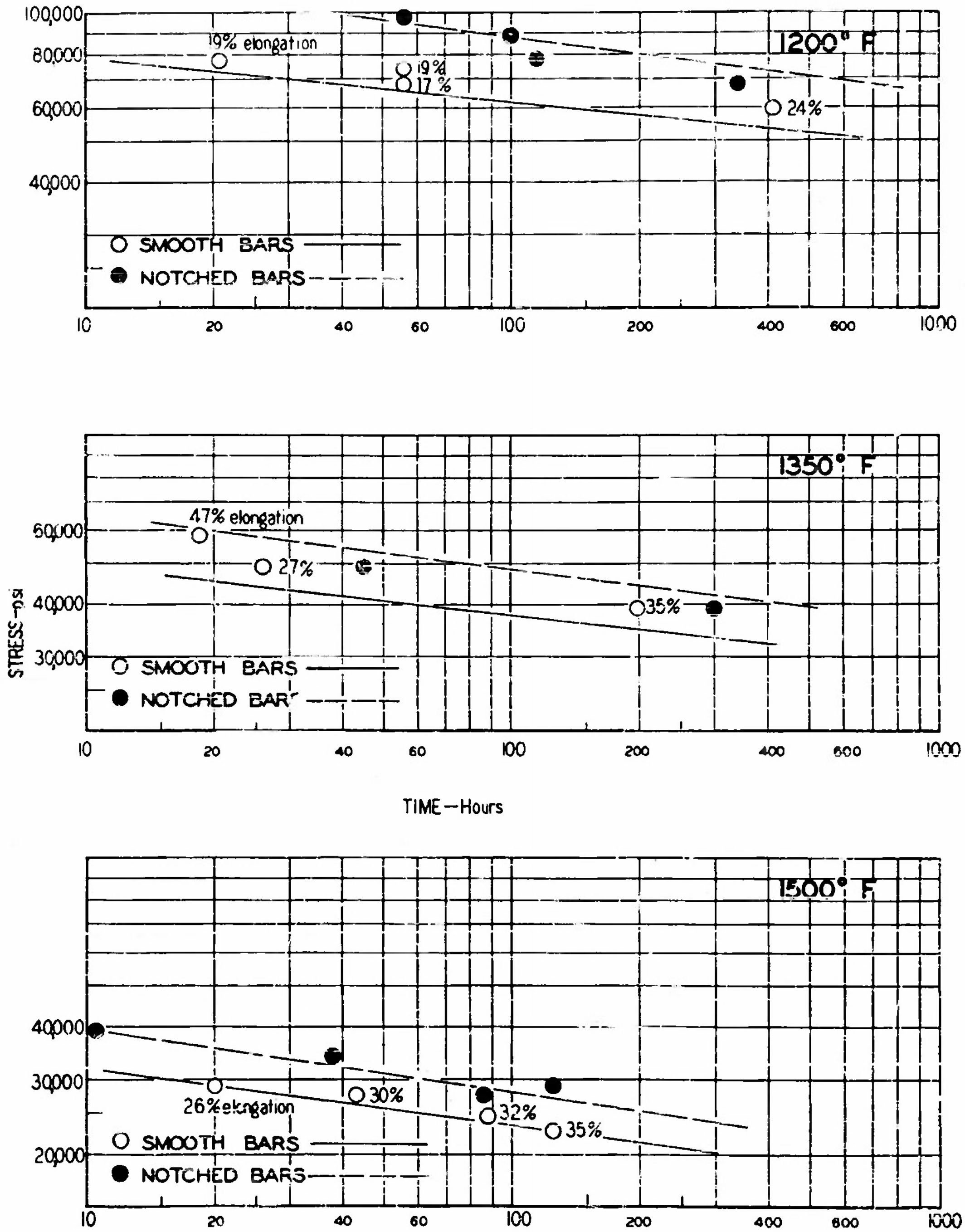


FIG.1 STRESS-RUPTURE-TIME CURVES FOR SMOOTH AND NOTCHED BARS OF S-816 (THE SMOOTH-BAR CURVES ARE AVERAGE RESULTS FOR THESE AND OTHER TESTS ON S-816. ALL CURVES WERE TAKEN FROM A PARAMETER PLOT USED IN THE ORIGINAL REPORT.)

HEAT TREATMENT	NOTCH GEOMETRY				REF.
	D	d	r	ANGLE	
2150°F, 1 HR, WQ + 1400°F, 12 HR, AC	0.177	0.125	0.005	60°	2

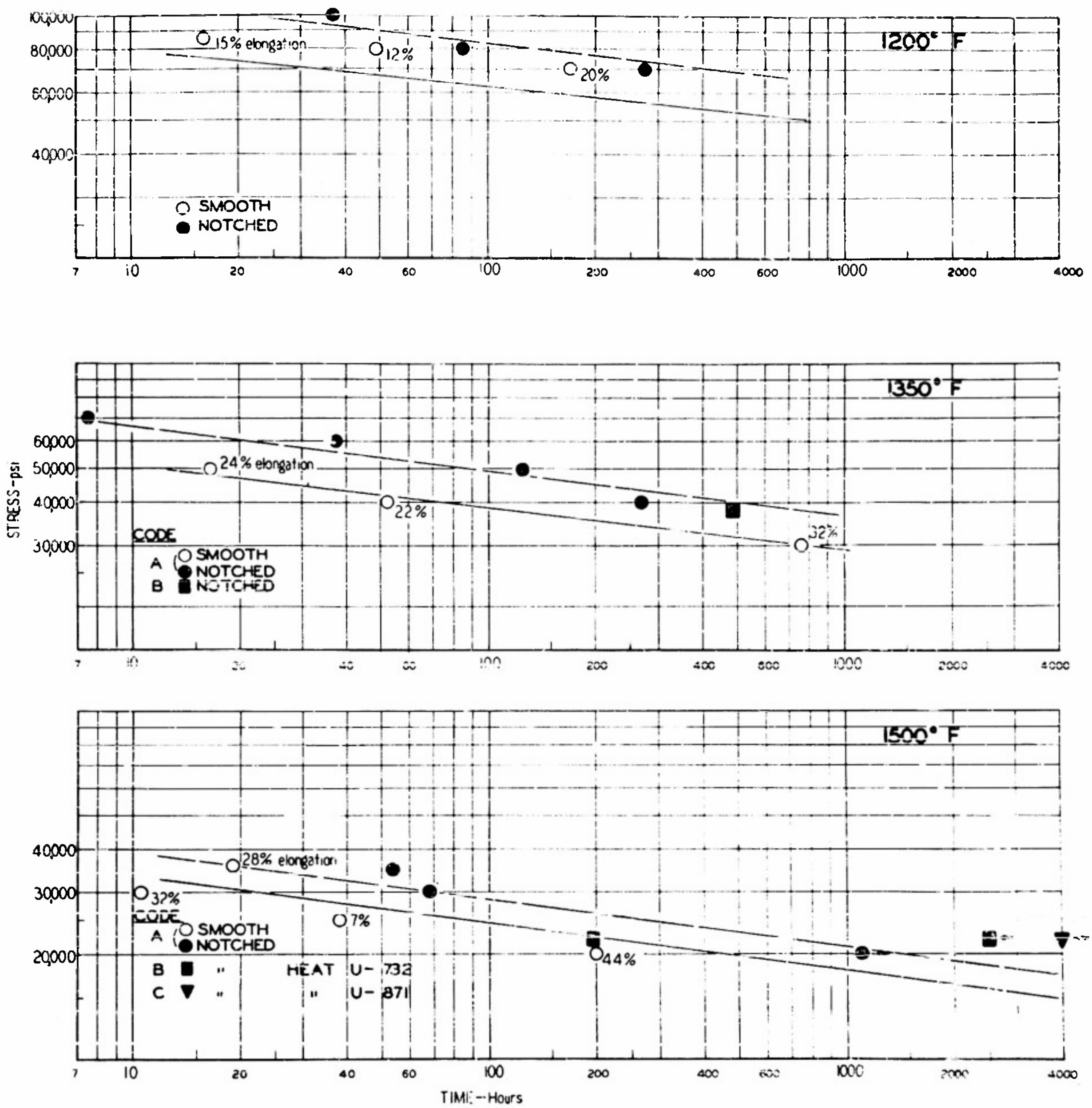


FIG. 2 STRESS-RUPTURE-TIME CURVES FOR SMOOTH AND NOTCHED BARS OF S-816-Cb+Ta. (THE SMOOTH-BAR CURVES ARE AVERAGE RESULTS FOR THESE AND OTHER TESTS ON S-816-Cb+Ta. ALL CURVES WERE TAKEN FROM A PARAMETER PLOT USED IN THE ORIGINAL REPORT.)

CODE	HEAT TREATMENT	NOTCH GEOMETRY				REF.
		D	d	r	ANGLE	
A	2150°F, 1 HR, W.Q.+ 1400°F, 12 HR, A.C.	0.177	0.125	0.005	60°	2
B, C	2250°F, 1 HR, W.Q.+ 1400°F, 16 HR, A.C.	0.275	0.1955	0.005	45°	3

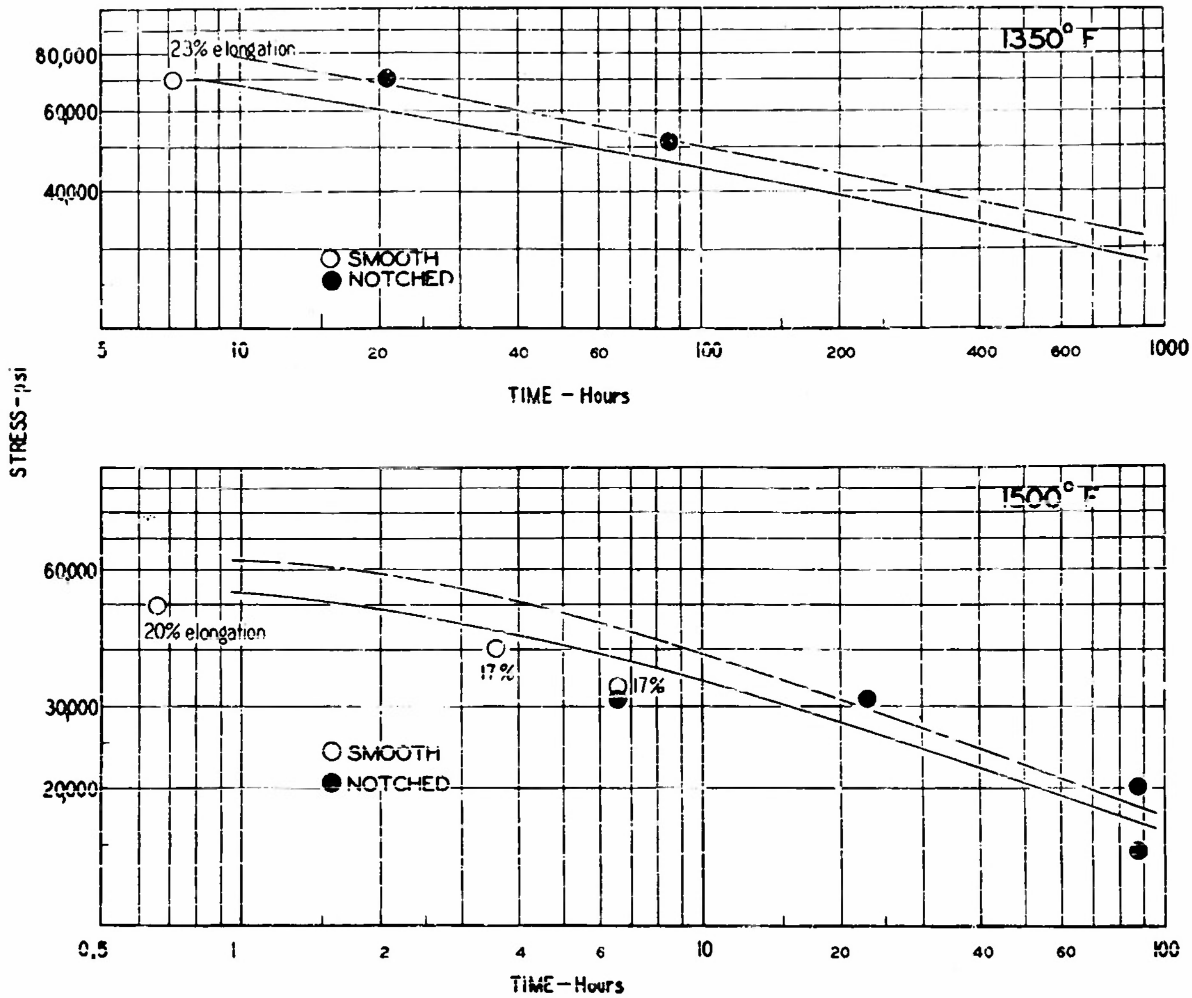


FIG 3 STRESS-RUPTURE-TIME CURVES FOR SMOOTH AND NOTCHED BARS OF M-252. (THE SMOOTH-BAR CURVES ARE AVERAGE RESULTS FOR THESE AND OTHER TESTS ON M252. ALL CURVES WERE TAKEN FROM A PARAMETER PLOT USED IN THE ORIGINAL REPORT.)

HEAT TREATMENT	NOTCH GEOMETRY D	d	r	ANGLE	REF.
FORGED FROM 2050°F + 1950°F, 4 HR, A.C. + 1650°F, 1 HR, F.C. TO 1000°F AT 90°F/HR.	0.177	0.125	0.005	60°	2

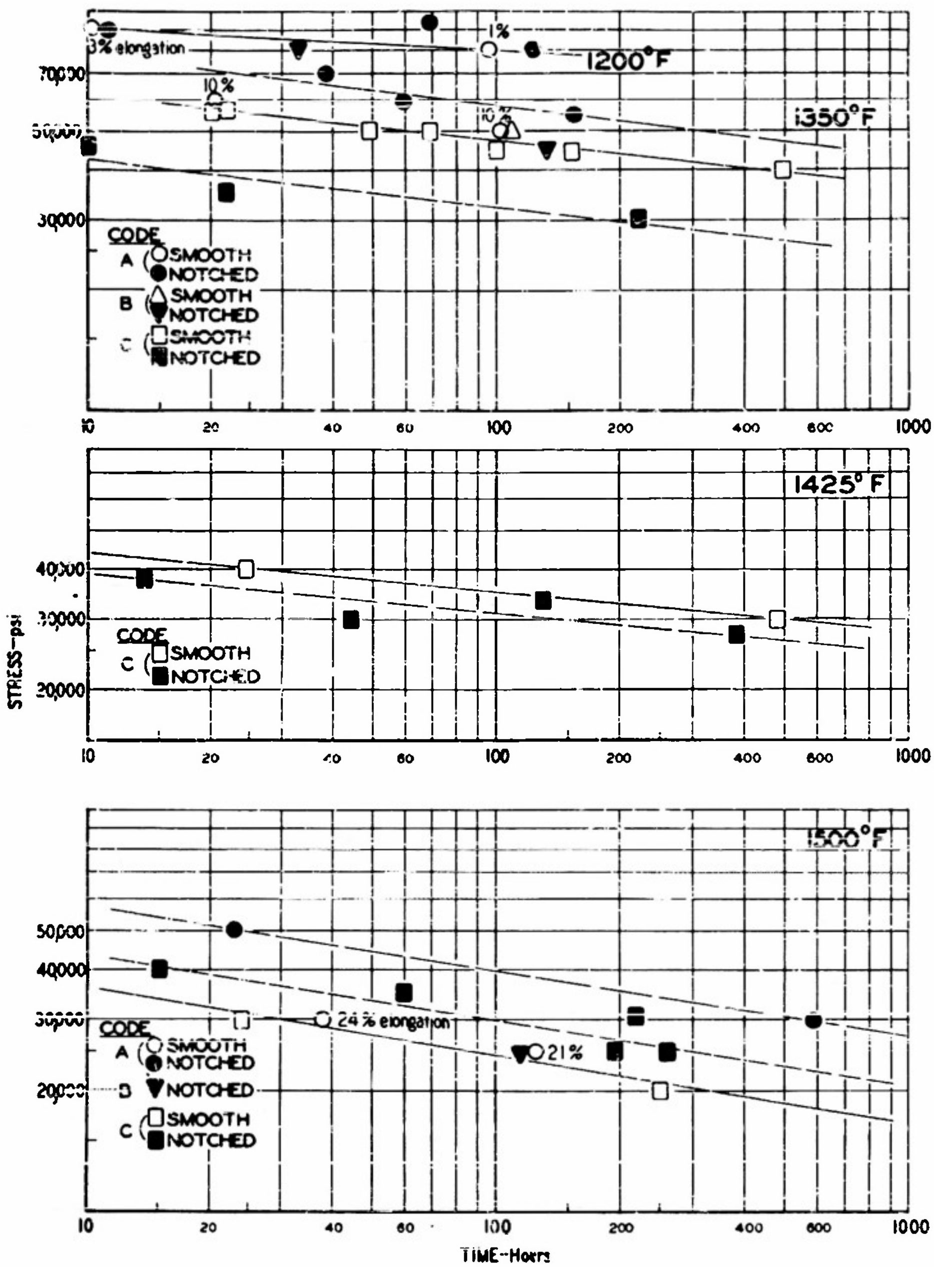


FIG.4 STRESS-RUPTURE-TIME CURVES FOR SMOOTH AND NOTCHED BARS OF INCONEL-X.

CODE	HEAT TREATMENT	NOTCH GEOMETRY				REF.
		D	d	r	ANGLE	
A	2100°F, 4 HR + 1550°F, 24 HR + 1300°F, 20 HR	0.370	0.250	0.005	60°	4
B	SAME AS ABOVE	SAME AS ABOVE, THEN GROUND FLAT ON OPPOSITE SIDES TO FORM A BAR 0.100-INCH THICK				4
C	2100°F, 4 HR, OIL + 1550°F, 24 HR, AC.+ 1300°F, 20 HR, AC.	0.424	0.300	NOT OVER 0.002	60°	5

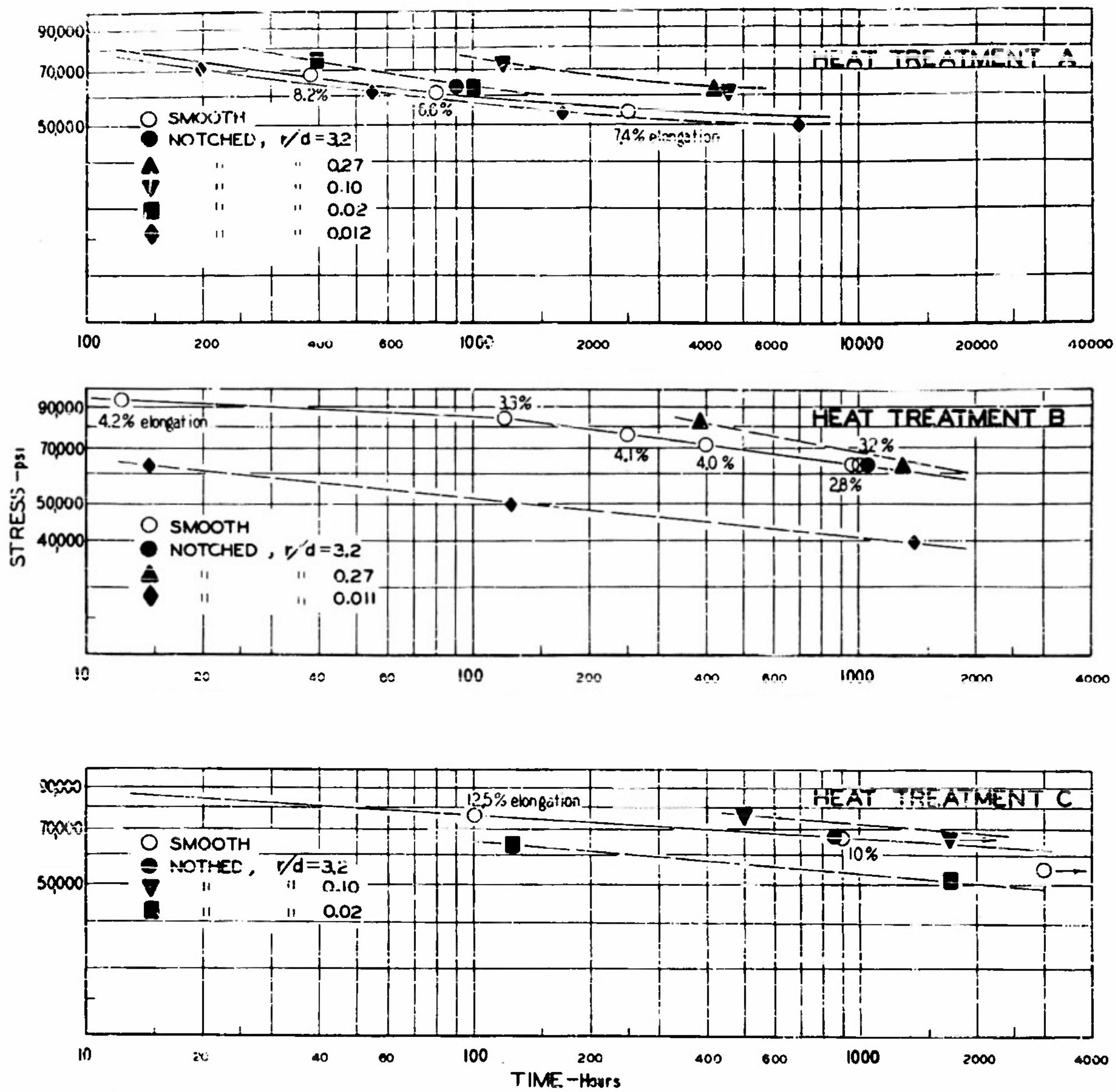


FIG. 5 STRESS-RUPTURE-TIME CURVES FOR SMOOTH AND NOTCHED BARS OF REFRACTALOY 26 AT 1200°F

	HEAT TREATMENT	HARDNESS DPH	A.S.T.M. GRAIN SIZE
A	1800°F, 20 MIN, OIL + 1500°F, 20 HR, AIR + 1200°F, 20 HR, AIR + 1500°F, 20 HR, AIR + 1200°F, 20 HR, AIR	330	7-8
B	1800°F, 20 MIN, OIL + 1350°F, 44 HR, AIR + 1200°F, 20 HR, AIR	375	7-8
C	2100°F, 1 HR, OIL + 1500°F, 20 HR, AIR + 1350°F, 20 HR, AIR + 1200°F, 20 HR, AIR	325	2-3

NOTCH GEOMETRY: NOTCH ROOT DIA. (d) = 15/32 IN.
NOTCH ANGLE = 60°
d/D = 0.75

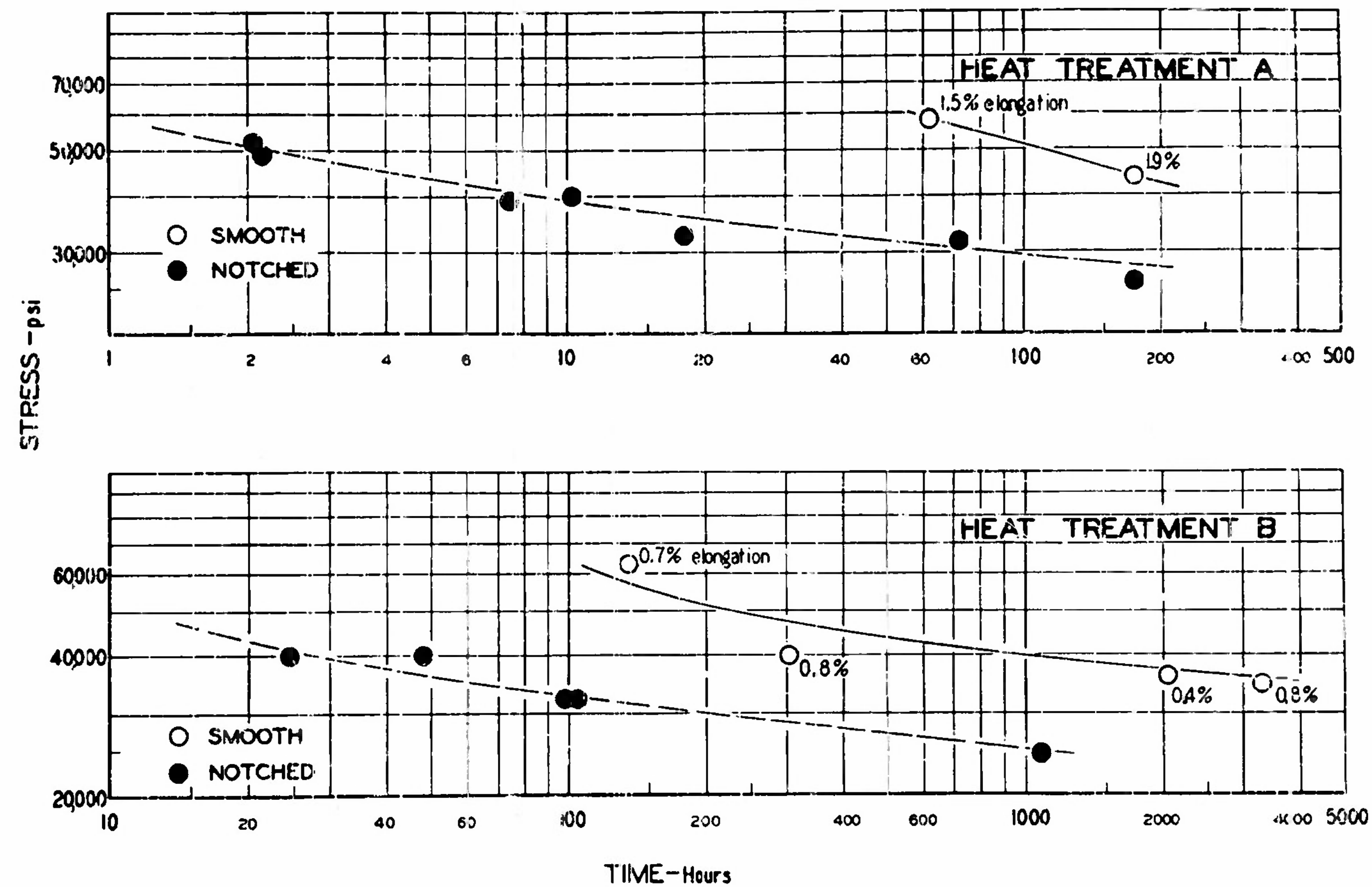


FIG.6 STRESS-RUPTURE-TIME CURVES FOR SMOOTH AND NOTCHED BARS OF K-42-B AT 1200°F.

	HEAT TREATMENT	HARDNESS DPH	A.S.T.M. GRAIN SIZE	REFERENCE
A	1750°F, 1HR, WQ + 1200°F, 24 HR, AC.	330	6-7	6
B	1950°F, 1HR, WQ + 1350°F, 20 HR, AC.	280	3-4	6

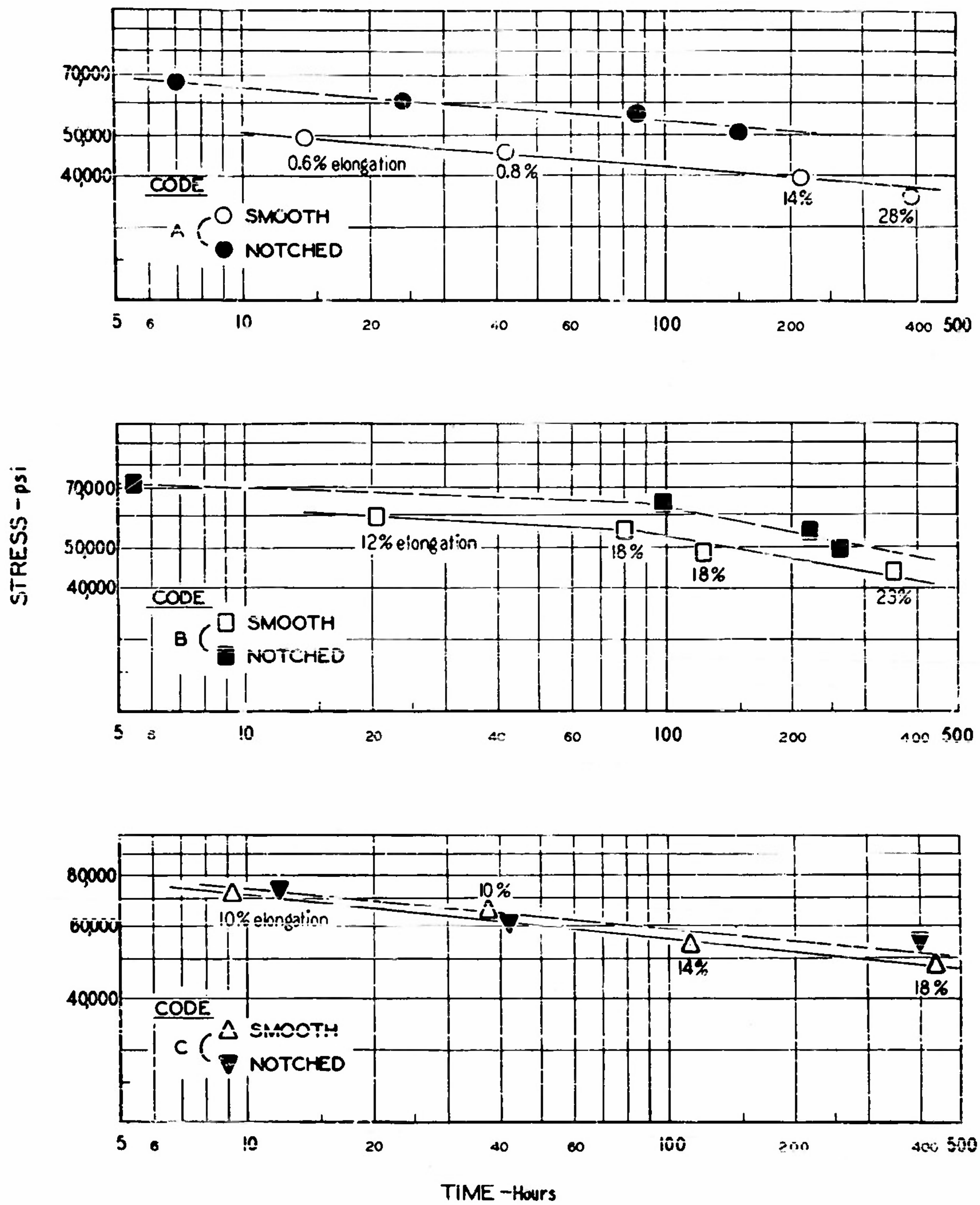


FIG.7 STRESS-RUPTURE-TIME CURVES FOR SMOOTH AND NOTCHED BARS OF 16-25-6 AT 1200° F

CODE	TREATMENT	HARDNESS (BHN)	REFERENCE
A	FORGED AT 1950°F + 2000°F, 1/4 HR, AC	187/200	2
B	FORGED AT 1950°F + COLD WORKED 20% AT 1350°F + 1200°F, 8 HR, AC	245/265	2
C	FORGED AT 1950°F + COLD WORKED 30% AT 1300°F + 1200°F, 8 HR, AC	280/320	2

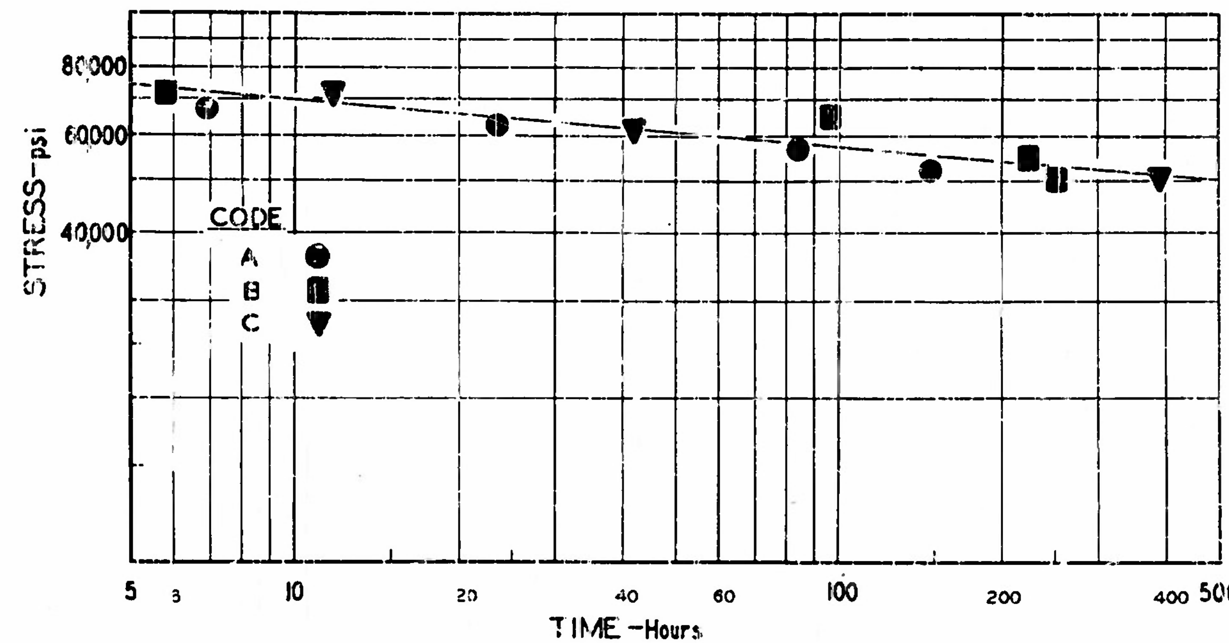


FIG.8 STRESS-RUPTURE-TIME CURVE FOR NOTCHED BARS OF 16-25-6 WITH DIFFERENT AMOUNTS OF PRIOR COLD WORKING.

CODE	TREATMENT	HARDNESS (BHN)	REFERENCE
A	FORGED AT 1950°F + 2000°F, 1/4 HR, AC.	187/200	2
B	FORGED AT 1950°F + COLD WORKED 20% AT 1350°F + 1200°F, 8 HR, A.C.	245/265	2
C	FORGED AT 1950°F + COLD WORKED 30% AT 1300°F + 1200°F, 8 HR, AC.	280/320	2

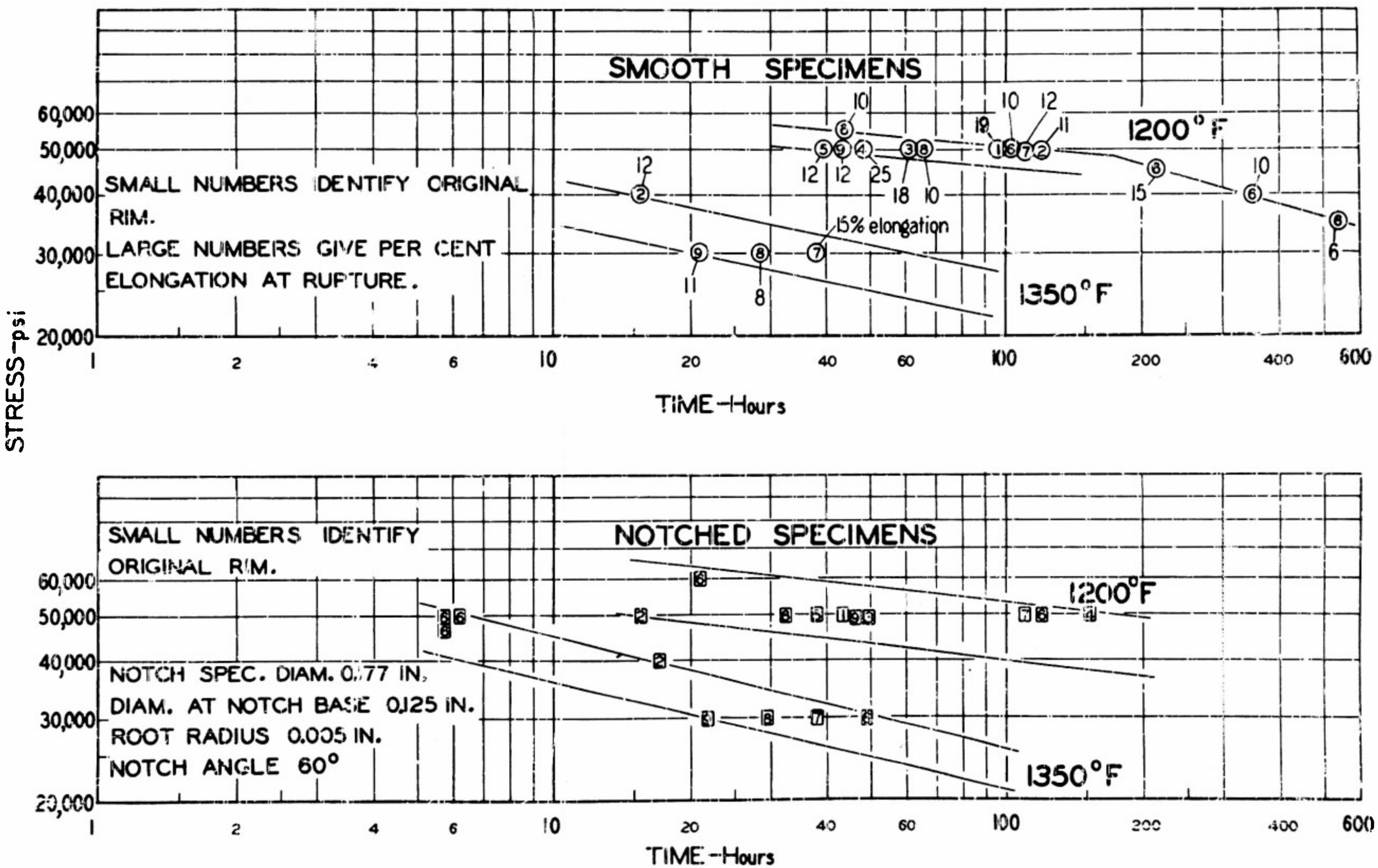


FIG. 9 STRESS-RUPTURE-TIME CURVES FOR SMOOTH AND NOTCHED SPECIMENS FROM COMMERCIAL COLD-WORKED 16-25-6 RIMS (DATA FROM THOMSON LABORATORY, GENERAL ELECTRIC COMPANY).

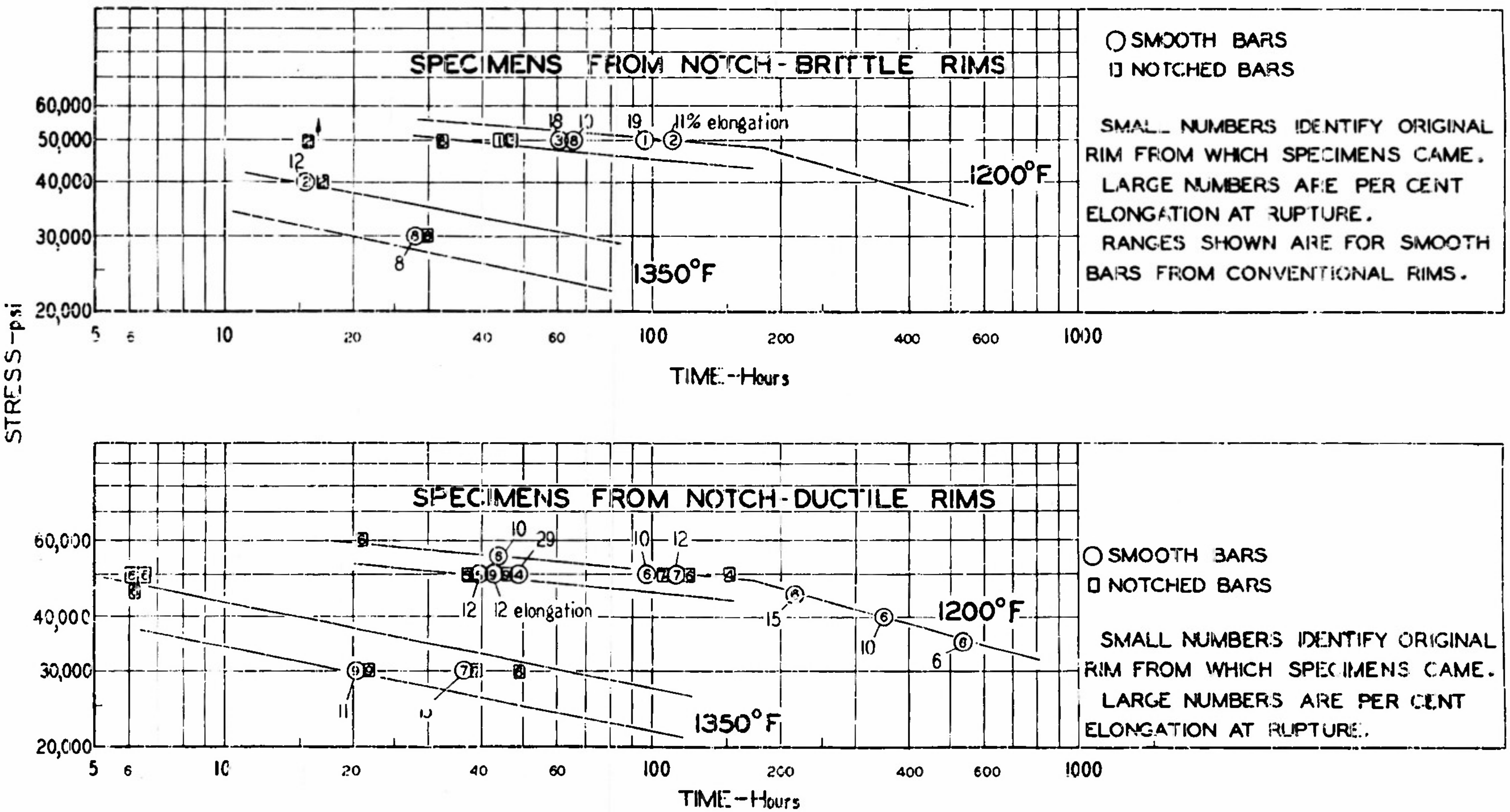


FIG.10 STRESS-RUPTURE-TIME DATA FOR 16-25-6 SPECIMENS FROM NOTCH-BRITTLE AND NOTCH-DUCTILE CONVENTIONAL COLD-WORKED RIMS (DATA FROM THOMSON LABORATORY, GENERAL ELECTRIC COMPANY).

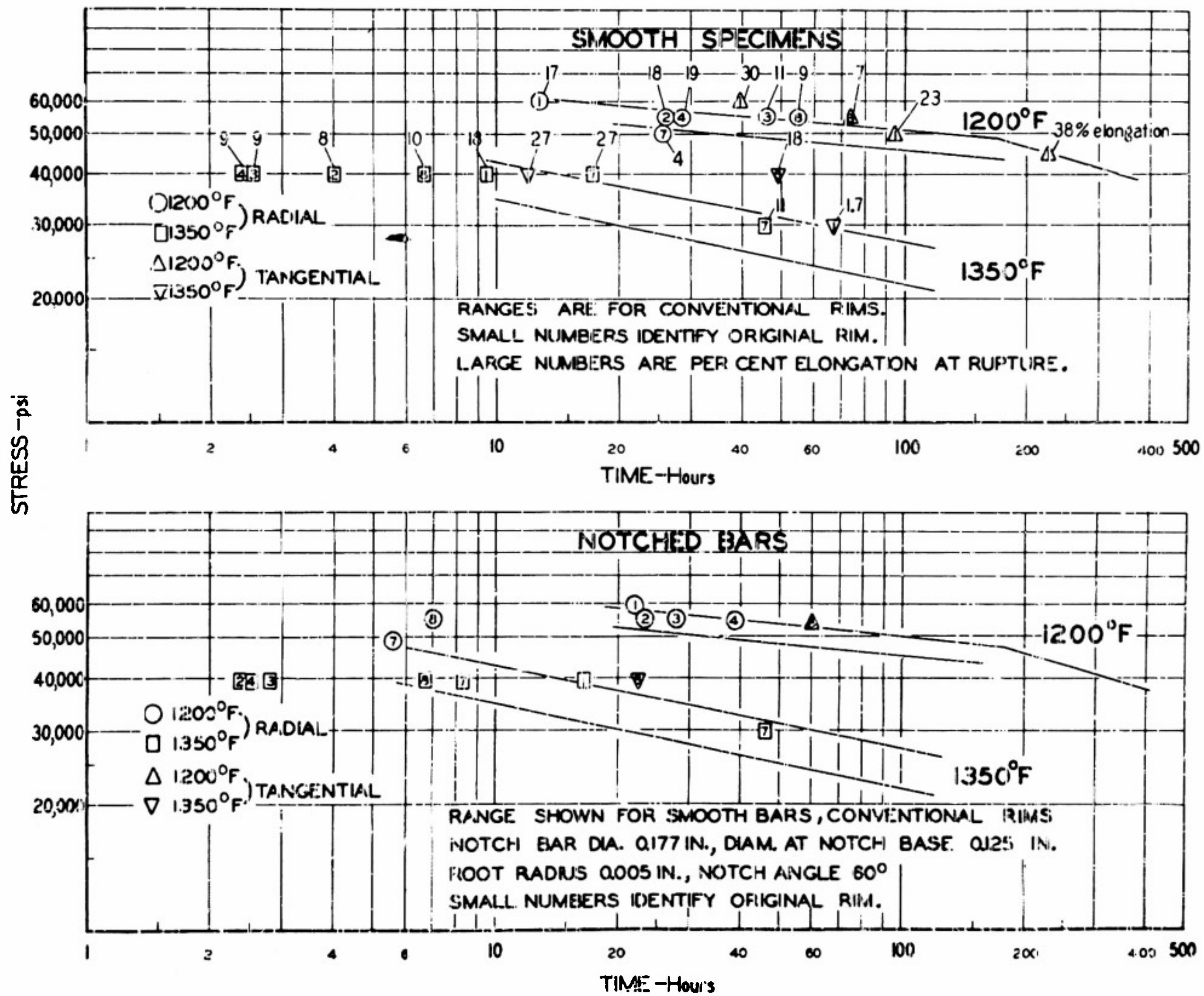


FIG.11 STRESS-RUPTURE-TIME CURVES FOR SMOOTH AND NOTCHED SPECIMENS FROM DIE EXPANDED RIMS OF 16-25-6. (DATA FROM THOMSON LABORATORY, GENERAL ELECTRIC COMPANY).

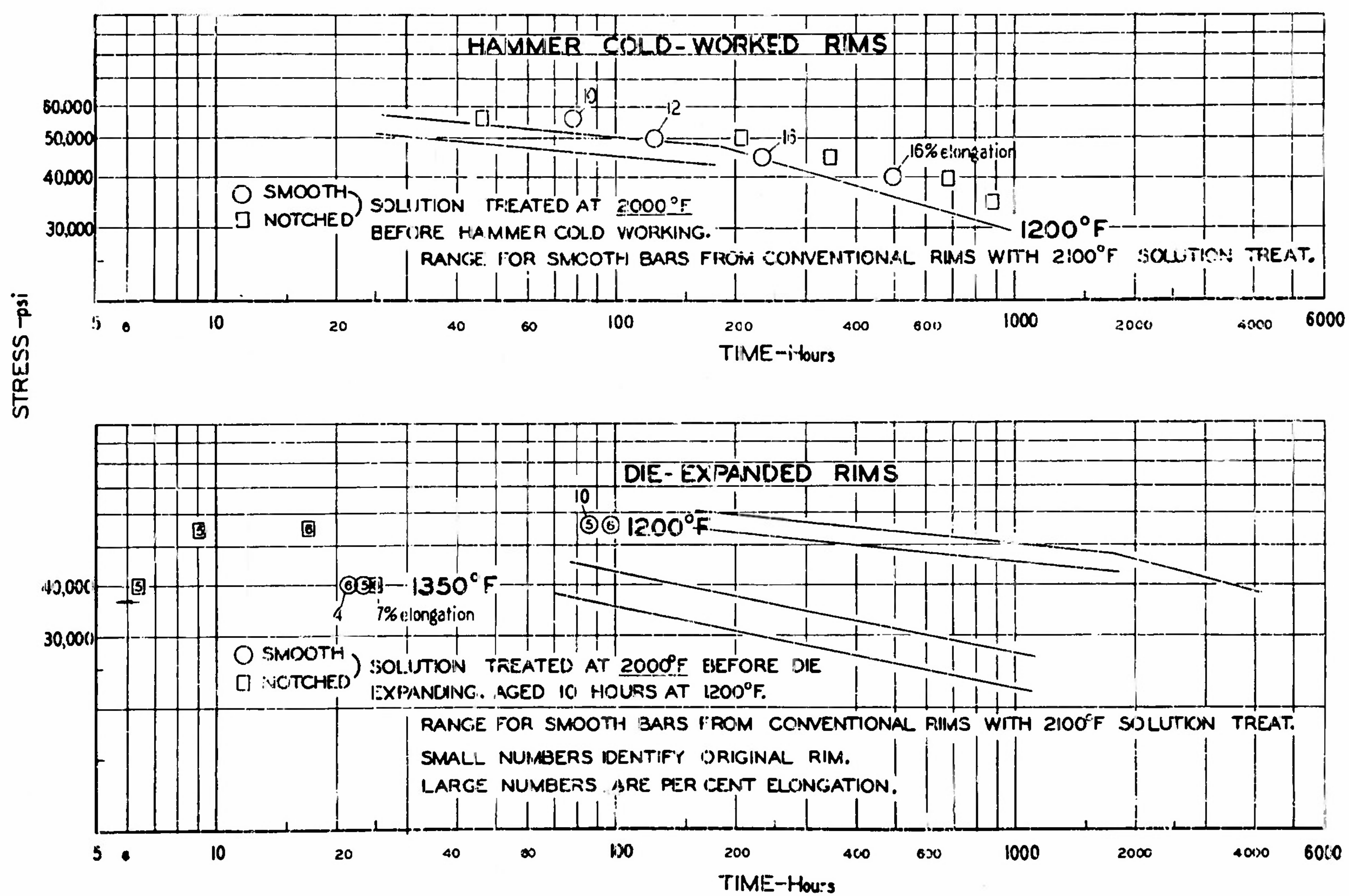


FIG.12 STRESS-RUPTURE-TIME PROPERTIES OF SMOOTH AND NOTCHED BARS FROM 16-25-6 RIMS SOLUTION-TREATED AT 2000°F AND 2100°F BEFORE COLD WORK. (DATA FROM THOMSON LABORATORY, GENERAL ELECTRIC COMPANY).

○ Smooth } Radial
● Notched }
△ Smooth } Tangential
▲ Notched }

Notch spec. dia. 0.177 in.
Dia. at notch base 0.125 in.
Root radius 0.005
Notch angle 60°

NOTE: NUMBERS NEAR POINTS GIVE PERCENT ELONG. AT RUPTURE.

NOTE: THE RANGE OF VALUES FOR SMOOTH BARS FROM CONVENTIONAL RIMS IS GIVEN BY "R".

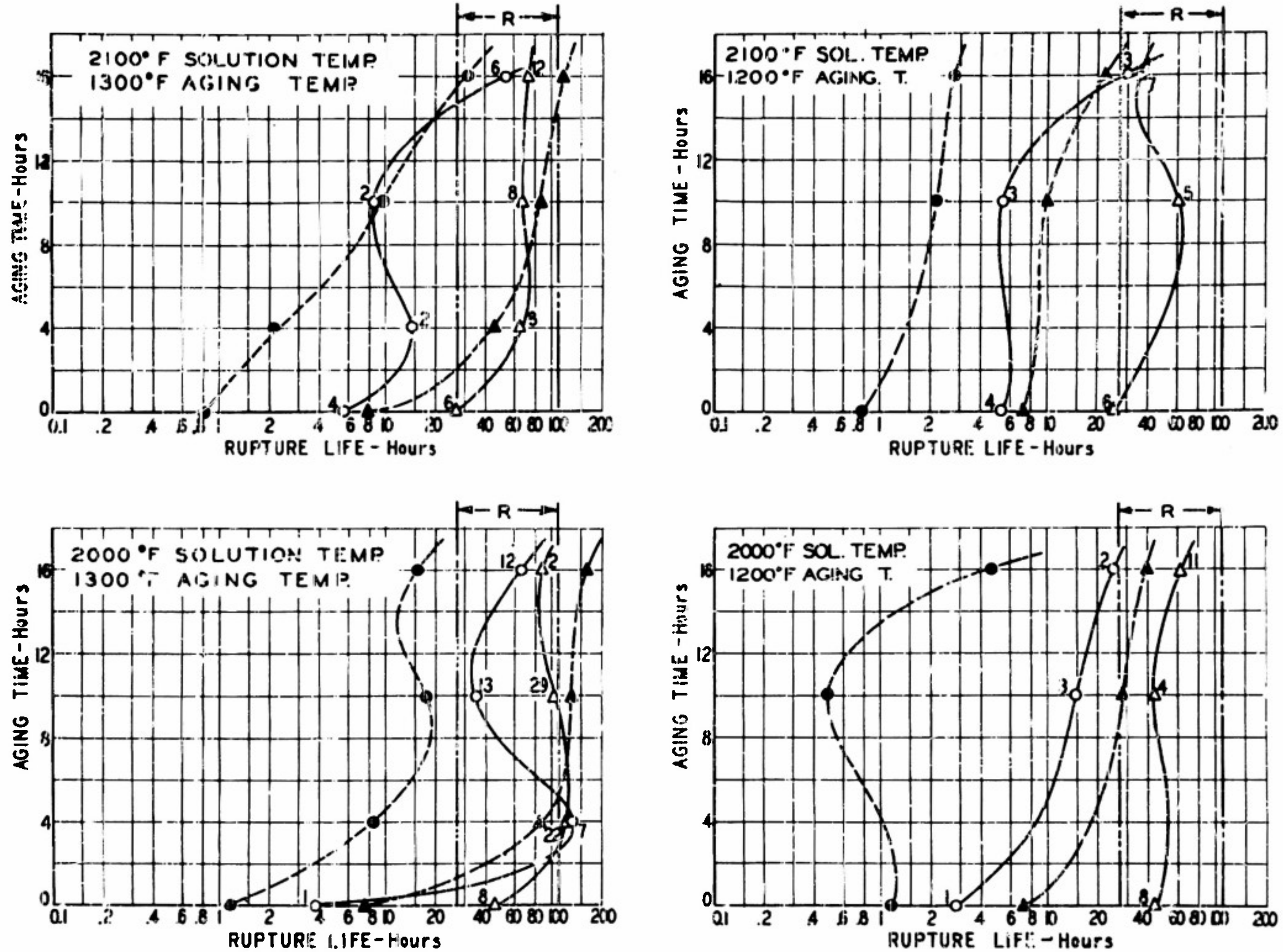
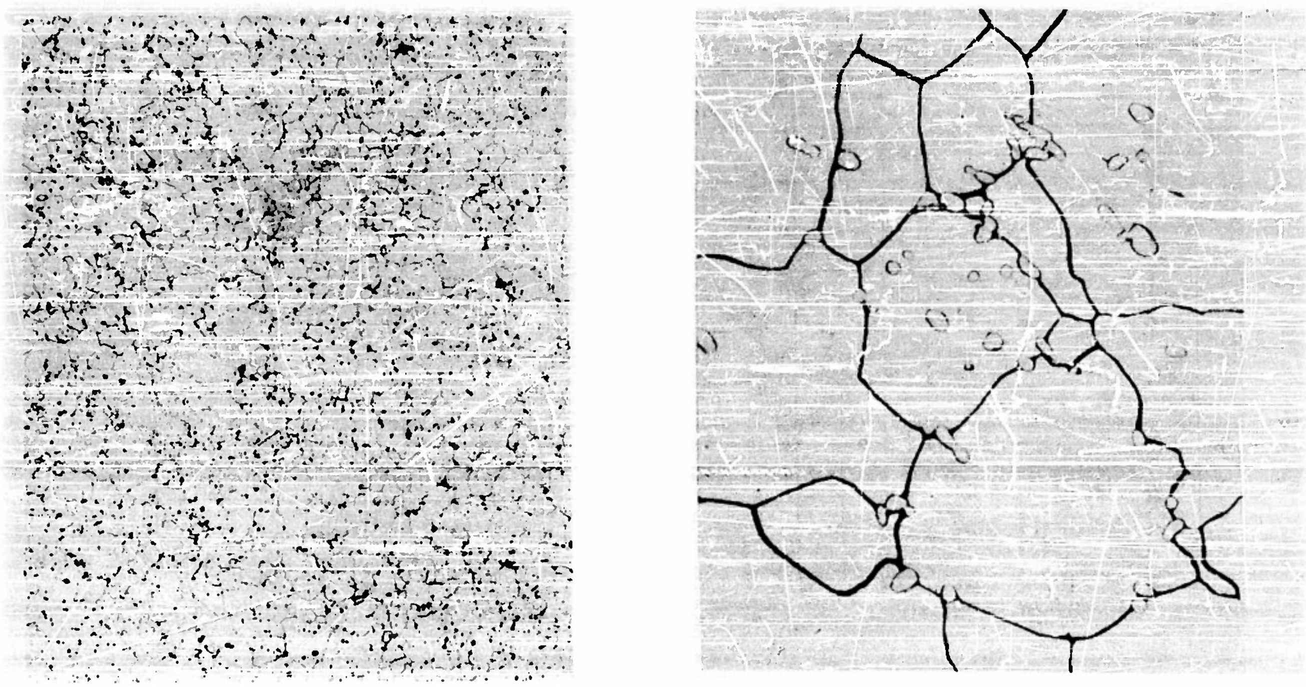
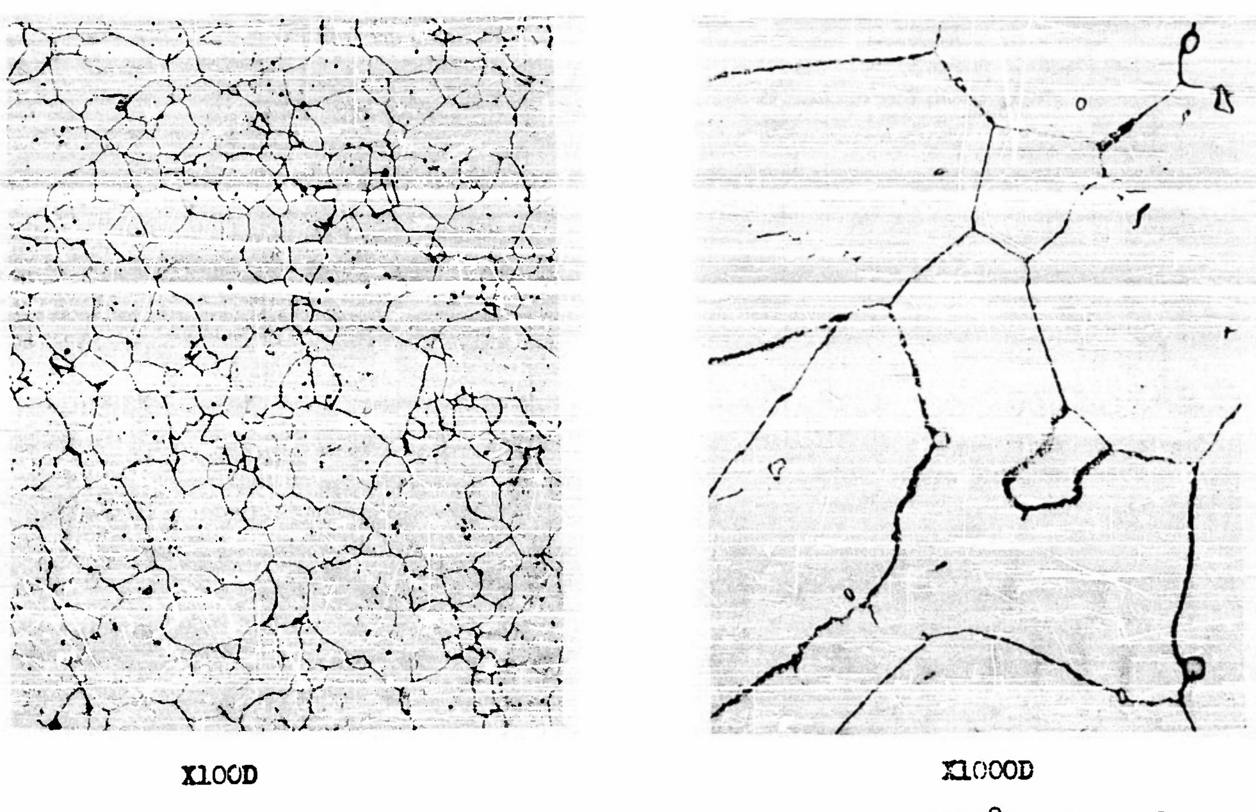


FIG. 13 EFFECT OF TIME OF AGING AT 1300°F AND 1200°F ON RUPTURE LIFE AT 1200°F AND 50,000 PSI OF 16-25-6 RIMS, DIE EXPANDED AFTER SOLUTION TREATMENT AT 2100° AND 2000° F.
(DATA FROM THOMSON LABORATORY — GENERAL ELECTRIC COMPANY.)



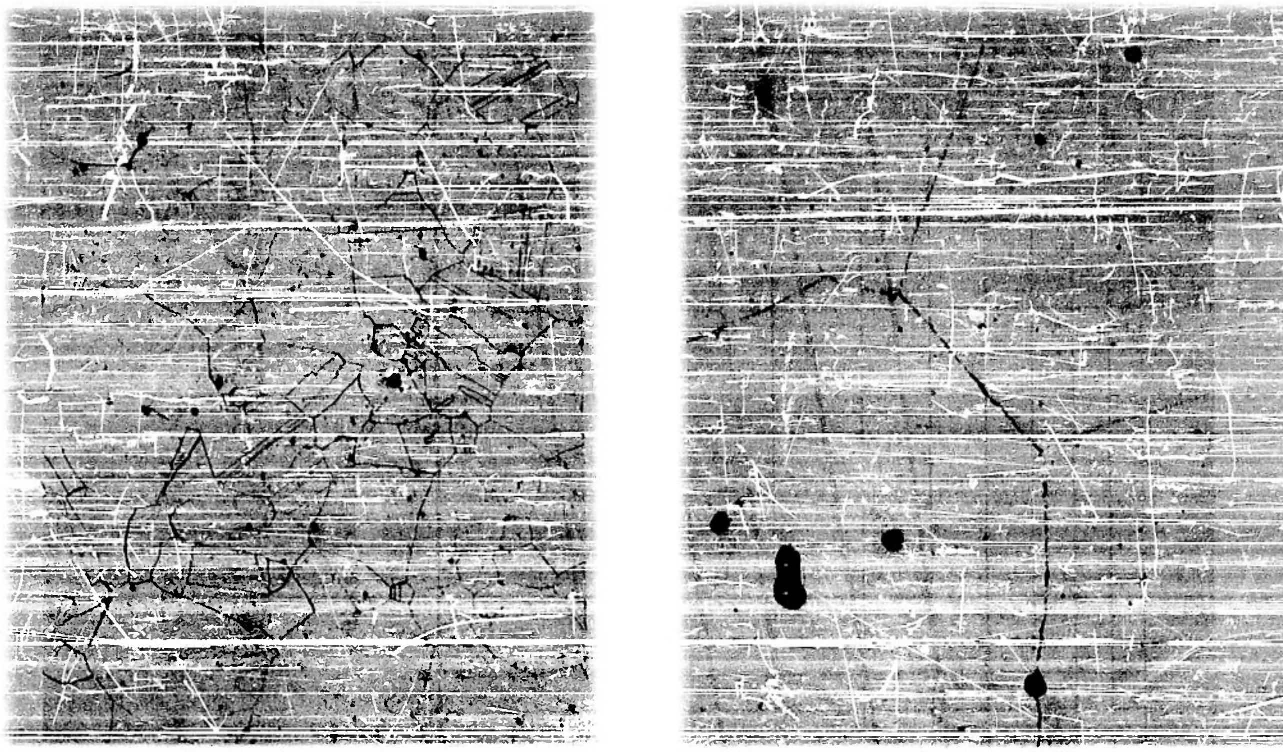
X100D X1000D

Figure 14. Original Microstructure of S-816 Alloy - 1 hour at 2150°F, water quenched + 12 hours at 1400°F, air cooled.



X100D X1000D

Figure 15. Original Microstructure of Waspaloy - 4 hours 1975°F, air cool + 4 hours at 1550°F, air cool + 16 hours at 1400°F, air cool.



X100D   X1000D

Figure 16. Original Microstructure of Inconel X-550 - 1 hour at 2150°F, air cool + 4 hours at 1600°F, air cool + 4 hours at 1350°F, air cool.

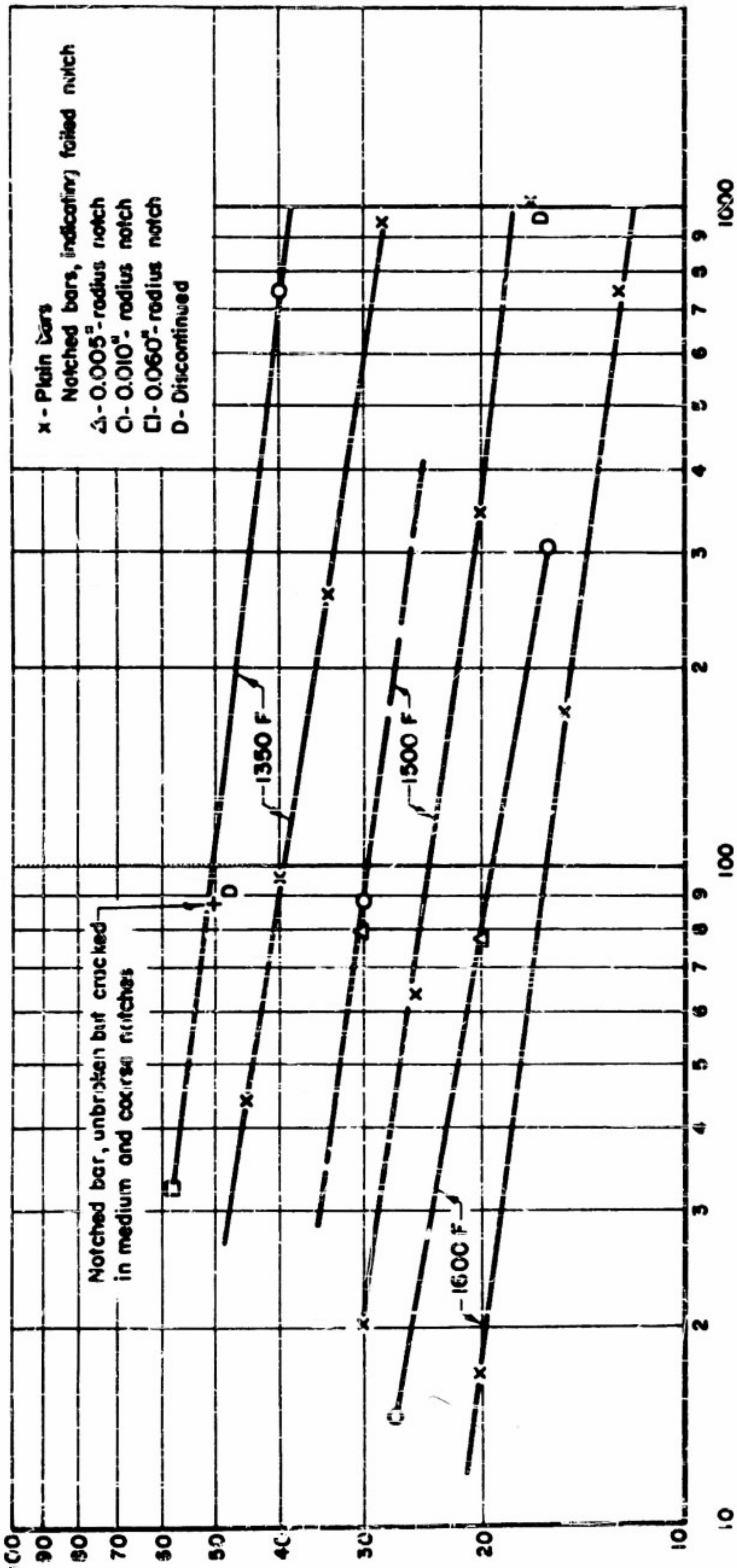


FIG. 17 STRESS-RUPTURE-TIME CURVES FOR SMOOTH (PLAIN) AND NOTCHED BARS OF S-816. (FROM CARLSON AND SIMMONS, REF. 9)

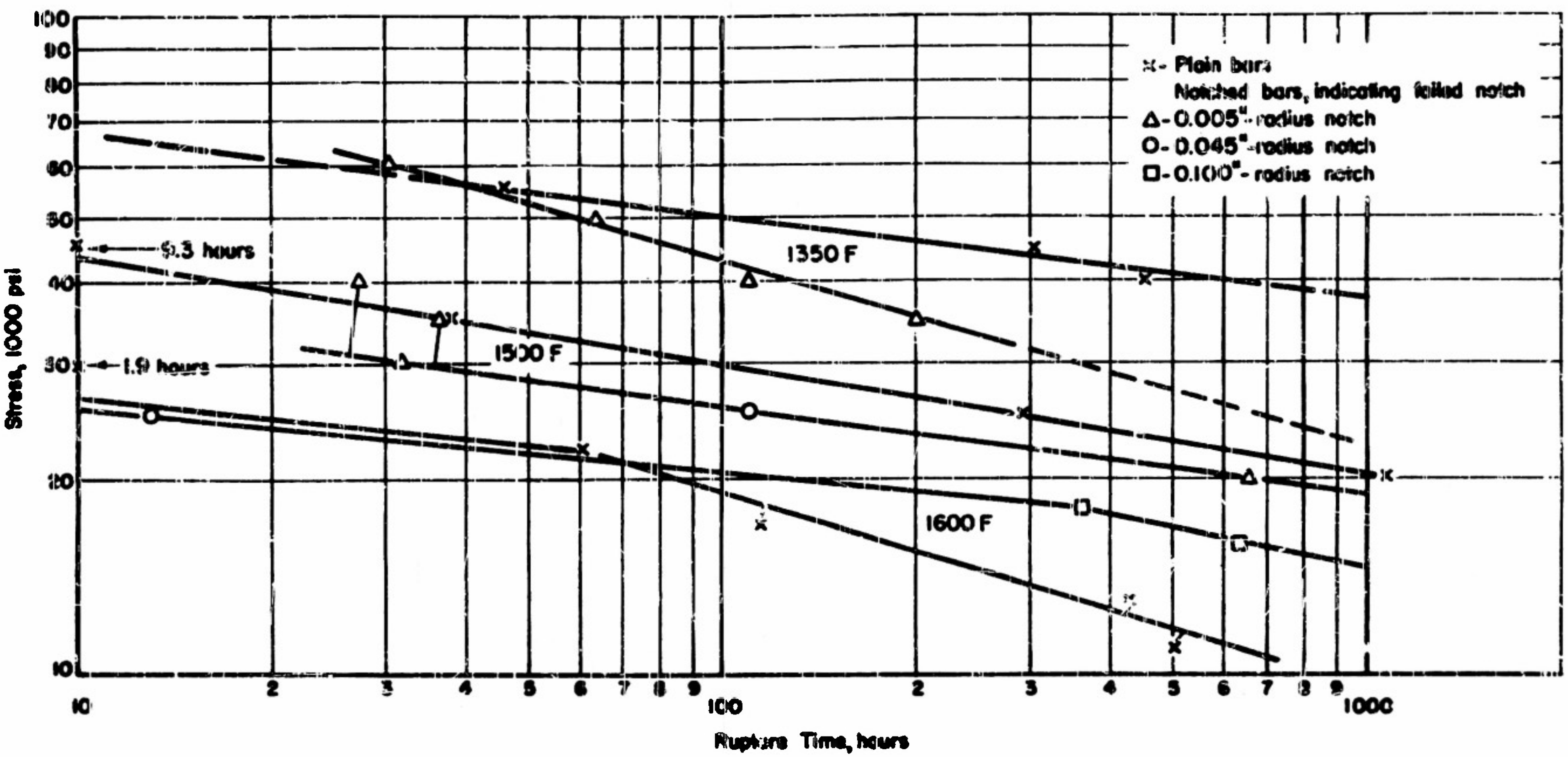


FIG. 18 STRESS-RUPTURE-TIME CURVES FOR SMOOTH (PLAIN) AND NOTCHED BARS OF INCONEL X-550. (FROM CARLSON AND SIMMONS, REF. 9)

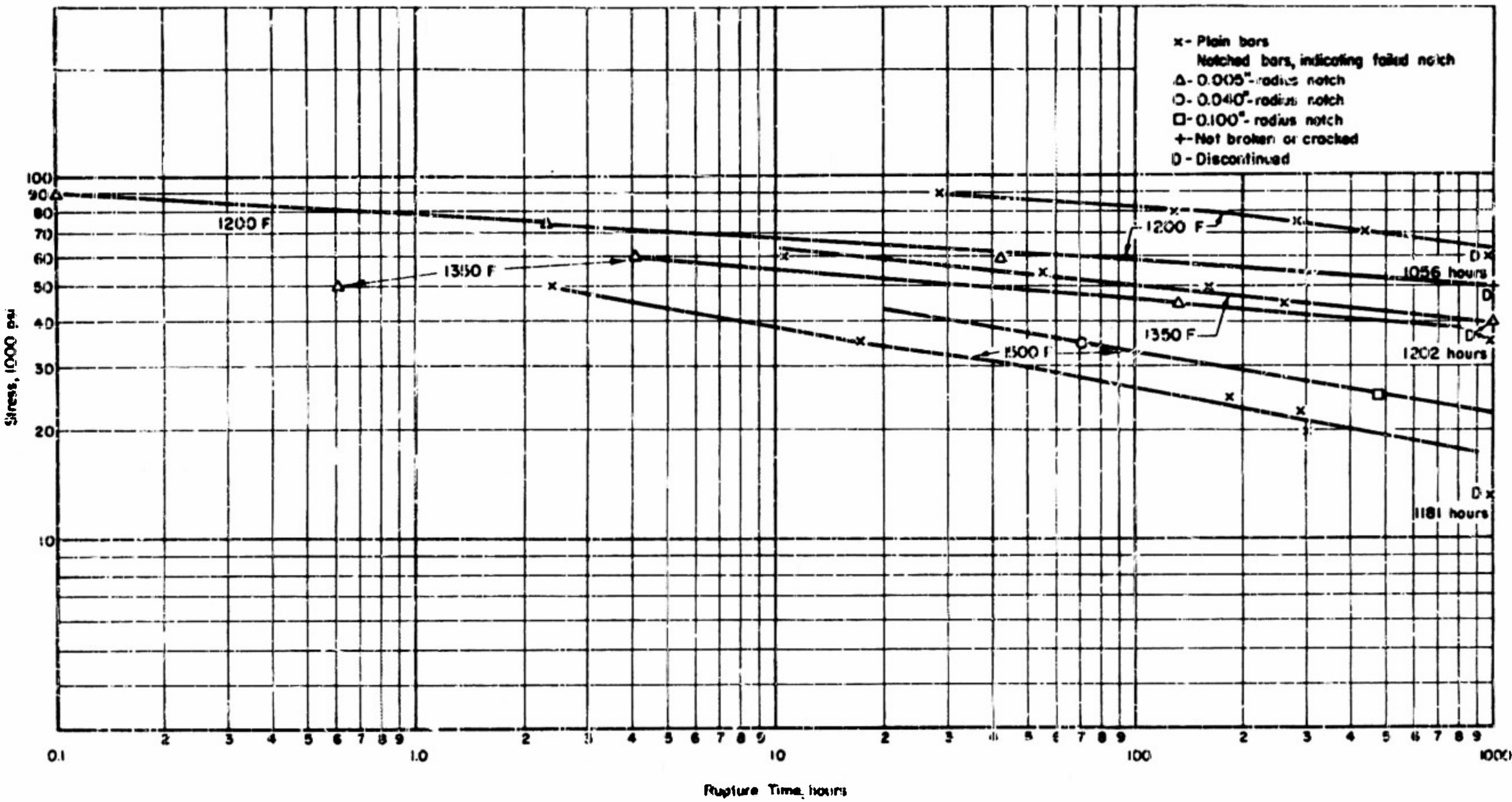


FIG. 19 STRESS-RUPTURE-TIME CURVES FOR SMOOTH (PLAIN) AND NOTCHED BARS OF WASPALOY. (FROM CARLSON AND SIMMONS, REF. 9)

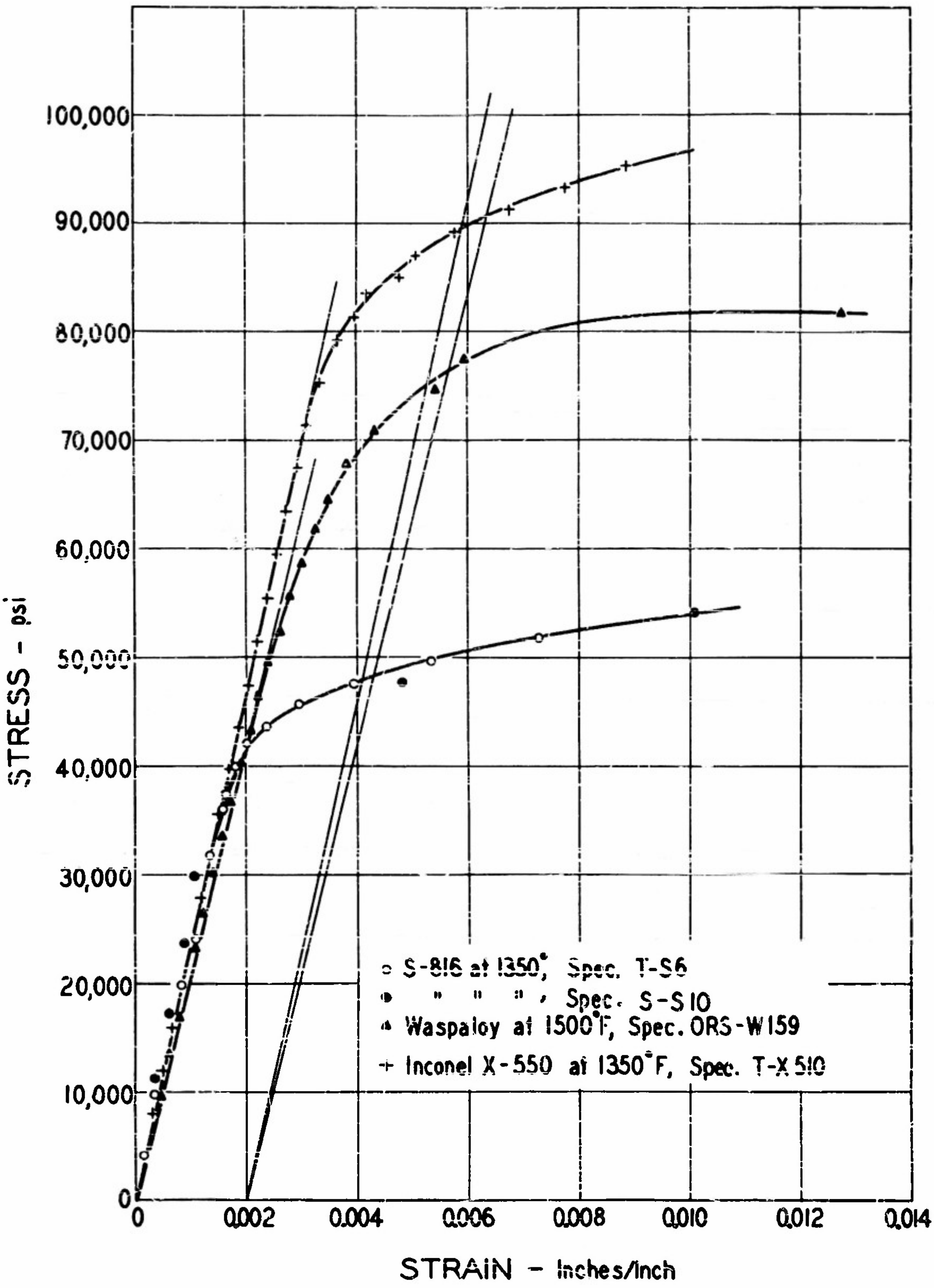


FIG. 20 SHORT-TIME TENSILE CHARACTERISTICS OF ALLOYS STUDIED.

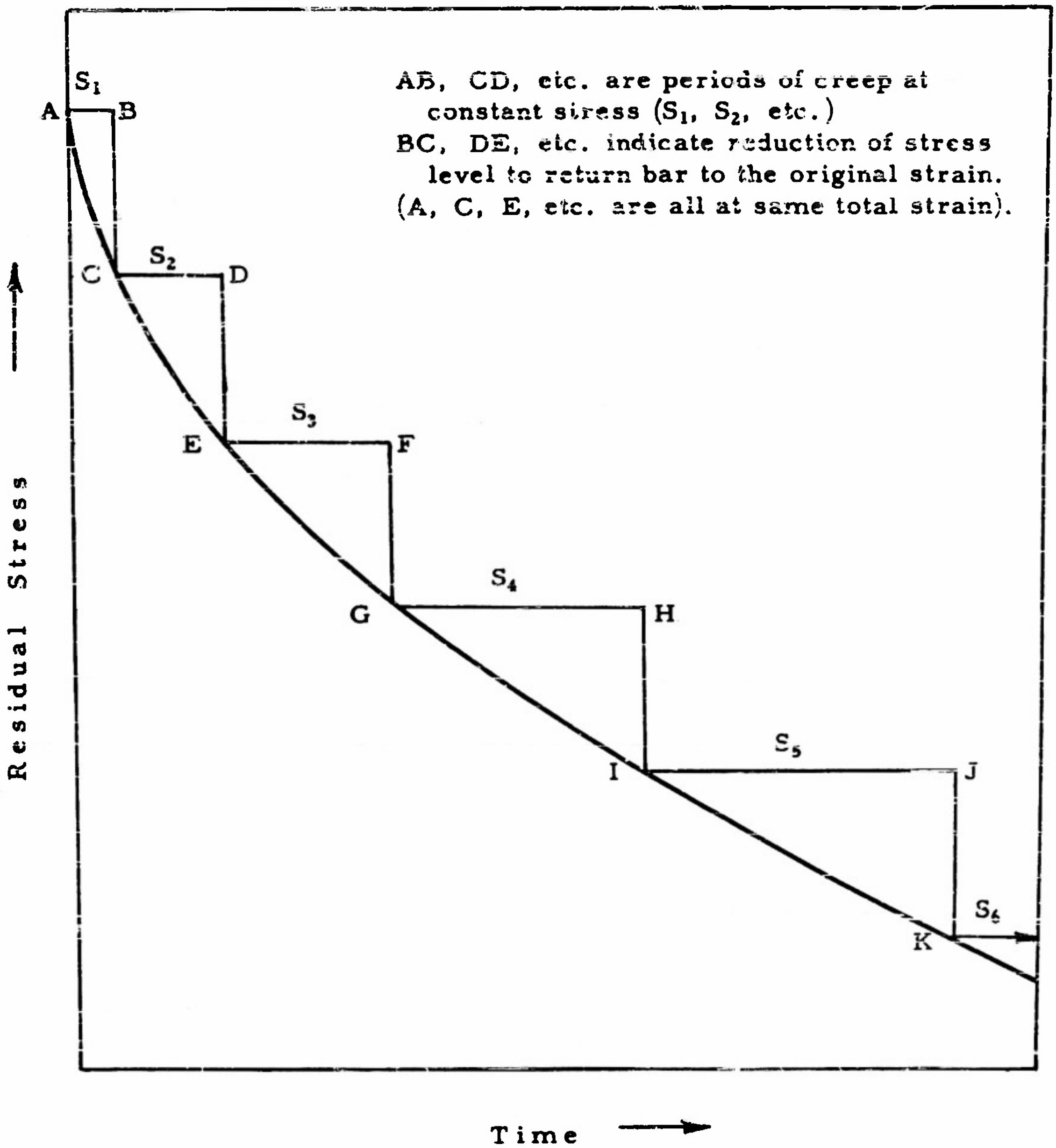


Fig. 21 STEP-WISE RELAXATION TEST OF A SMOOTH SPECIMEN IN PURE TENSION.

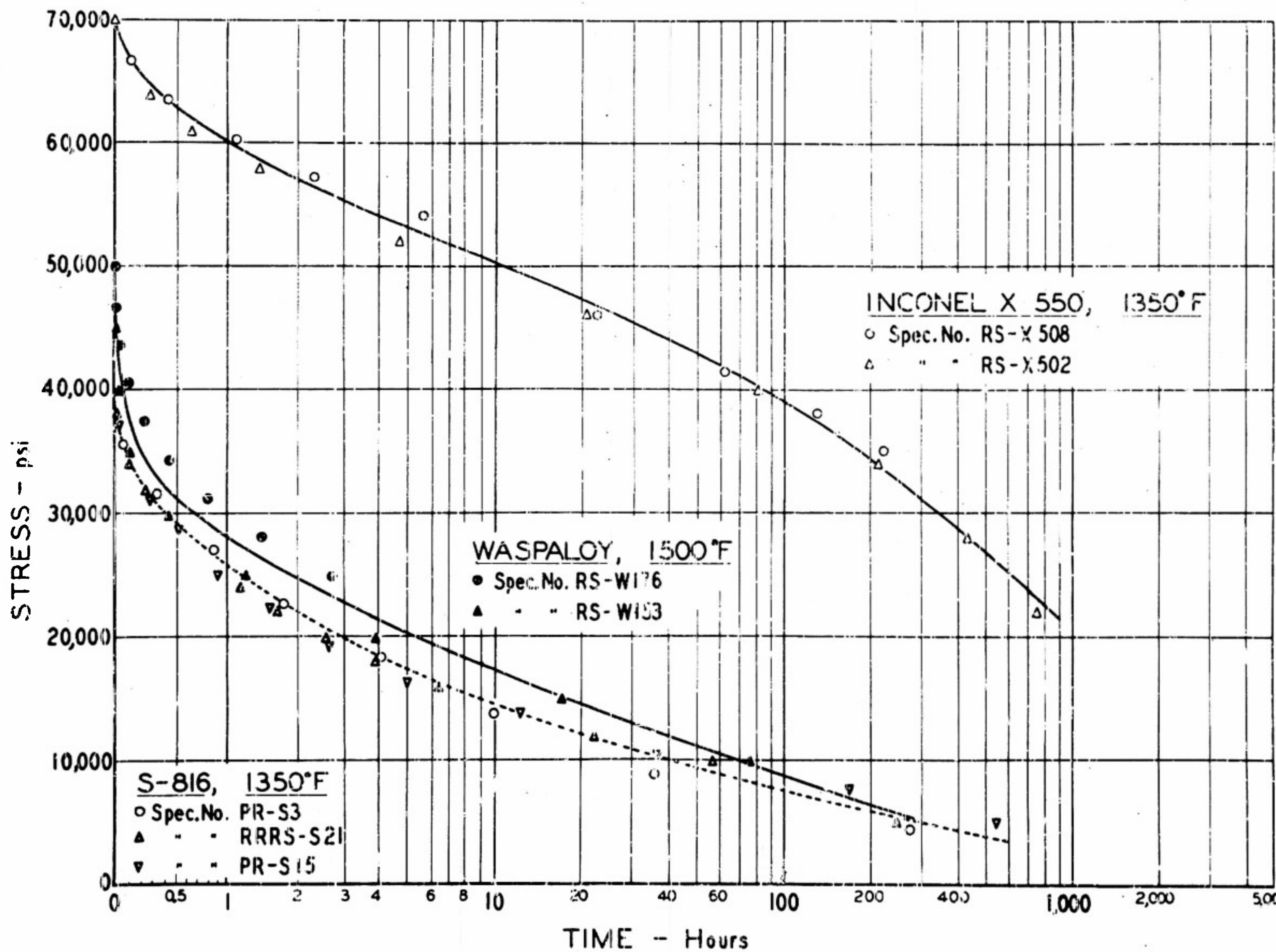


FIG. 22 REPRODUCIBILITY OF RELAXATION TEST DATA.

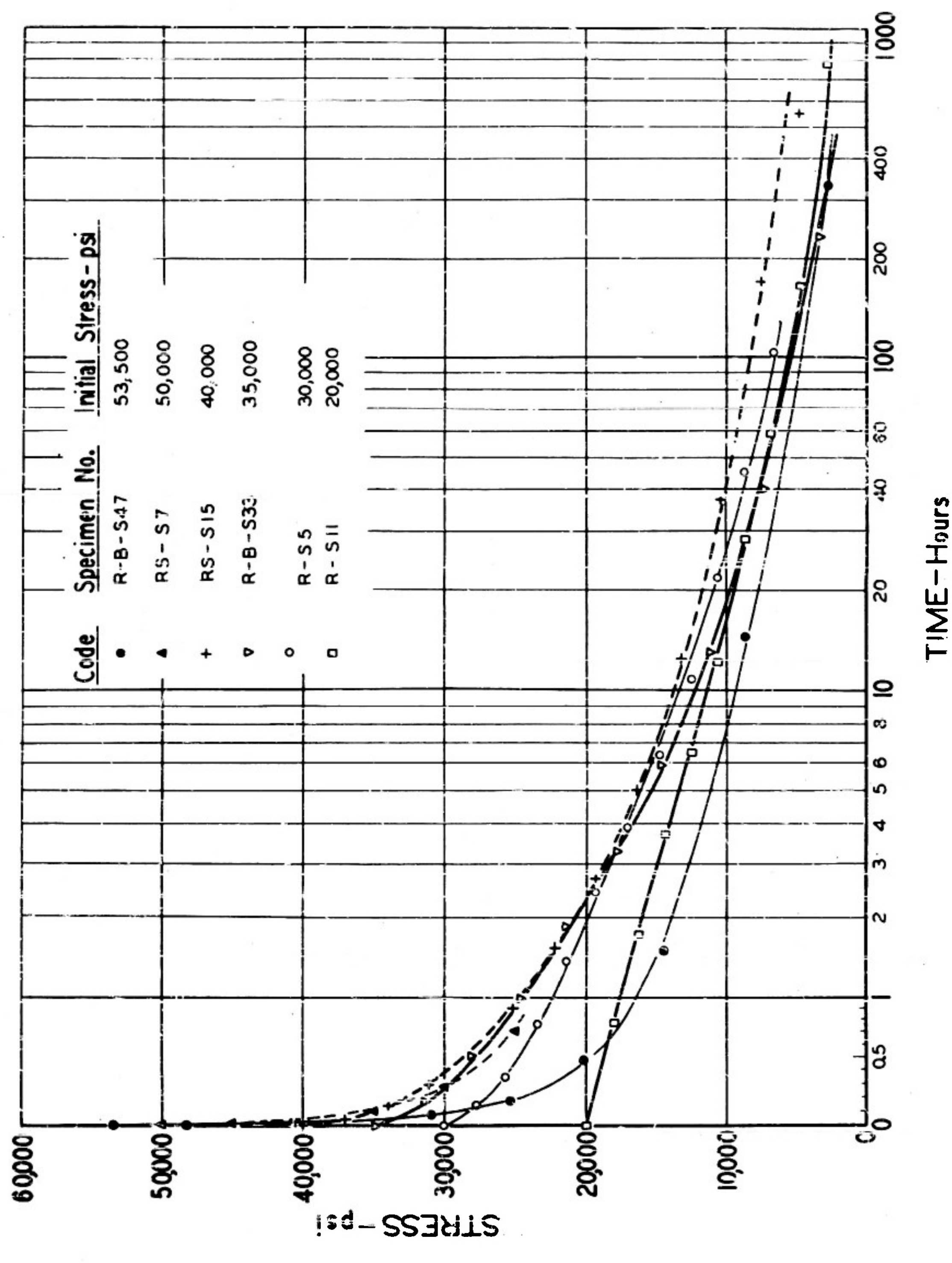


FIG. 23 RELAXATION CHARACTERISTICS OF S-816 AT 1350°F.

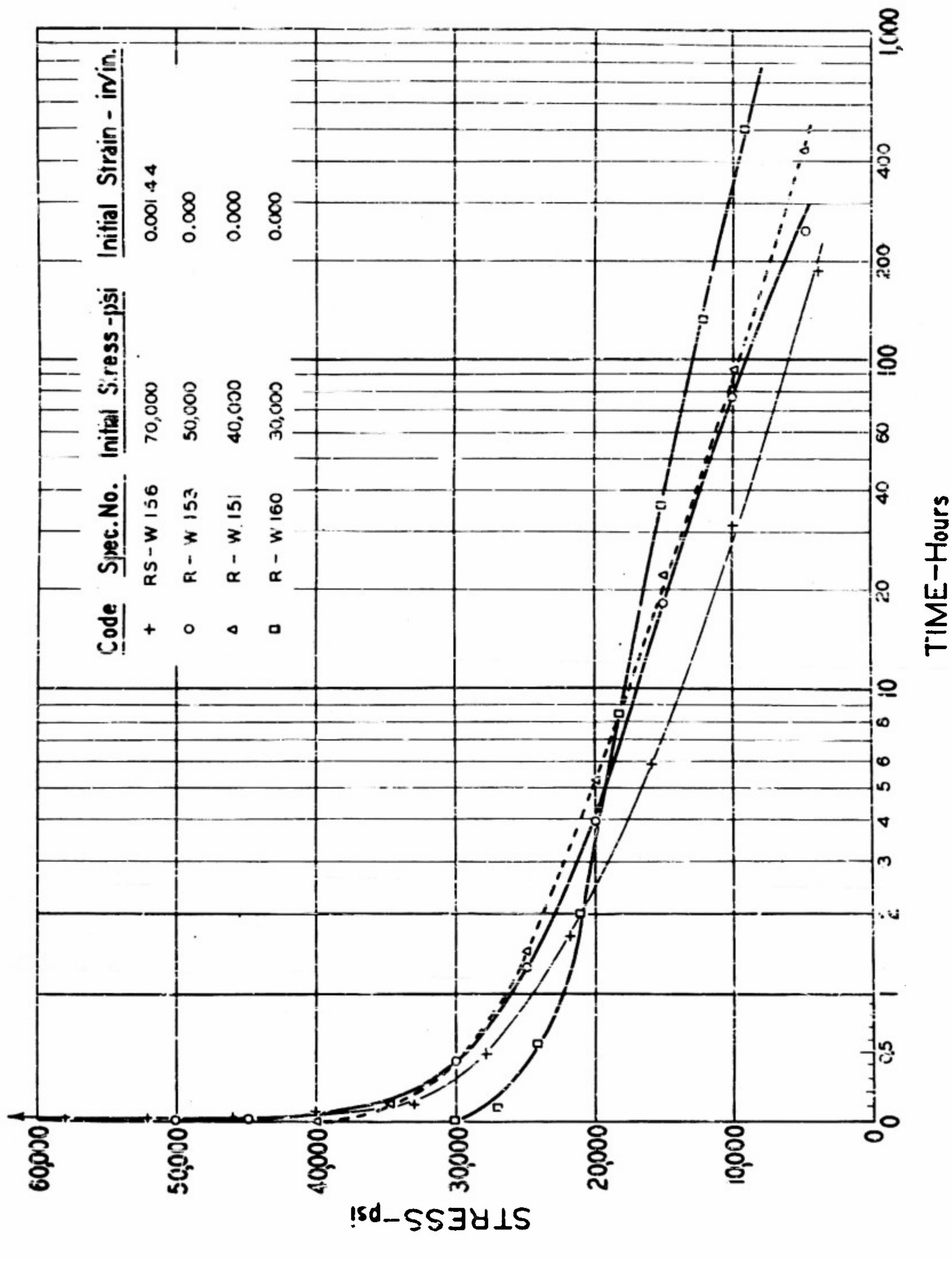


FIG.24 RELAXATION CHARACTERISTICS OF WASPALOY AT 1500°F

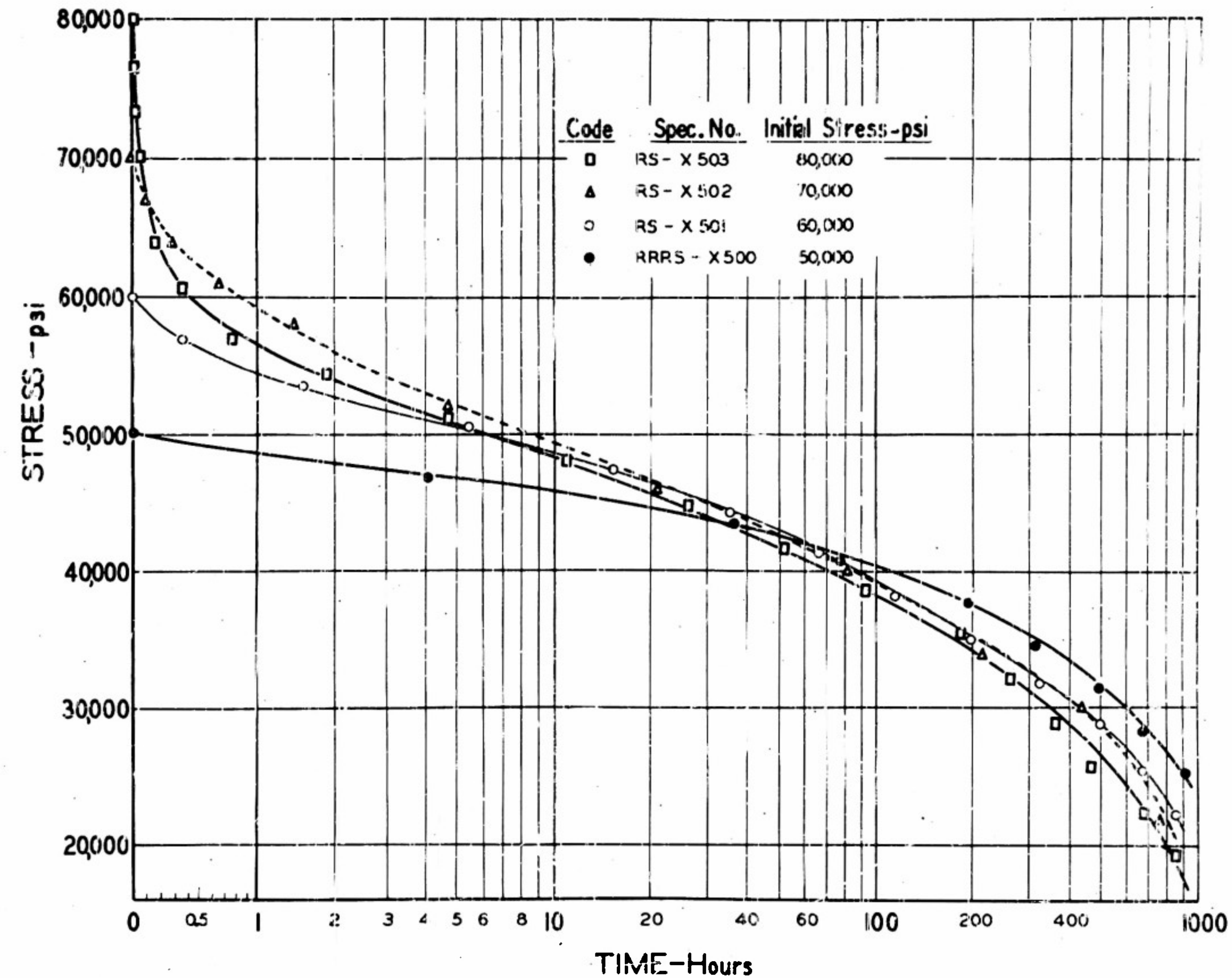


FIG.25 RELAXATION CHARACTERISTICS OF INCONEL X-550 AT 1350°F

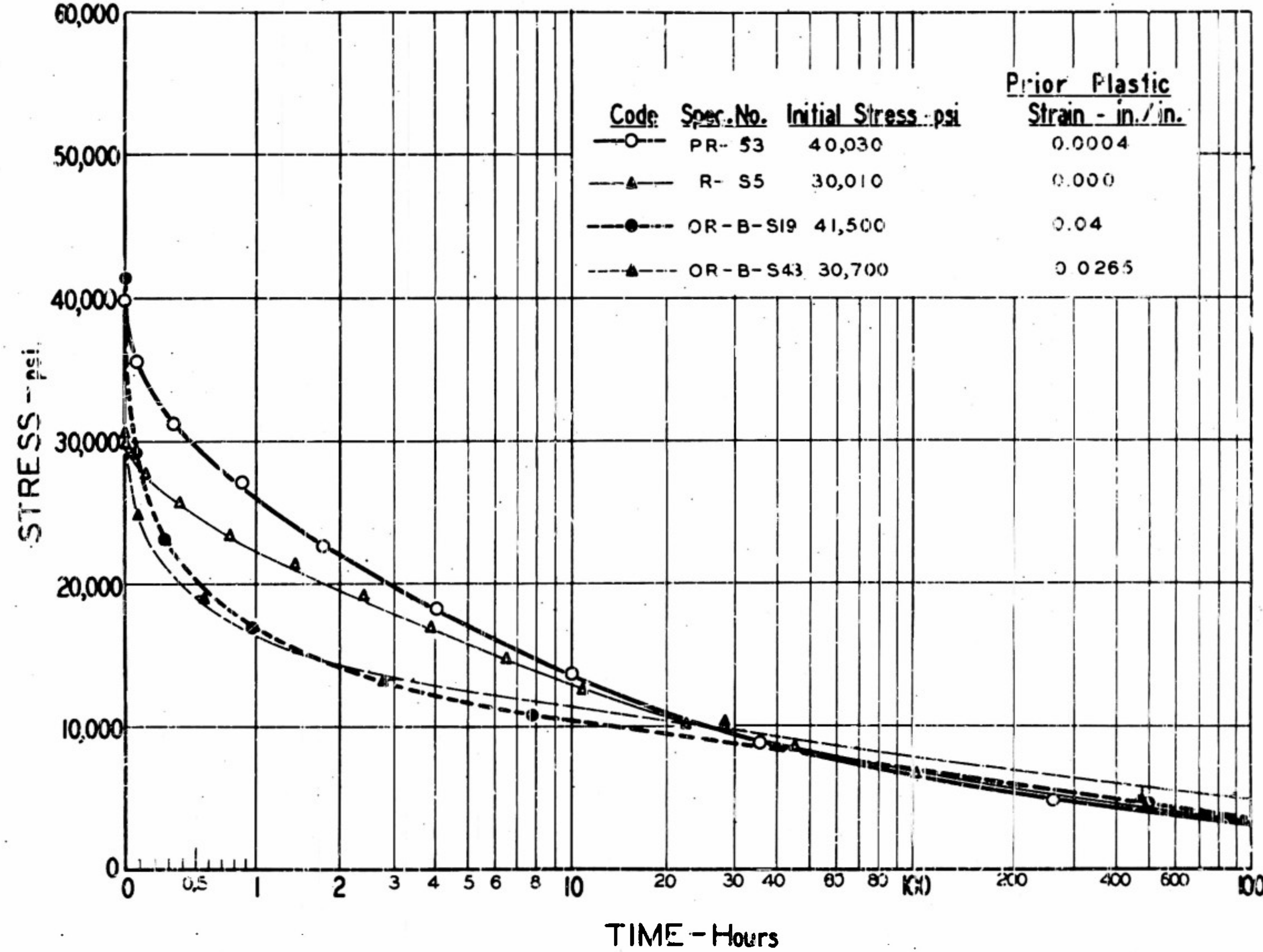


FIG. 26 EFFECT OF PRIOR PLASTIC STRAIN ON RELAXATION OF S-816 AT 1350° F

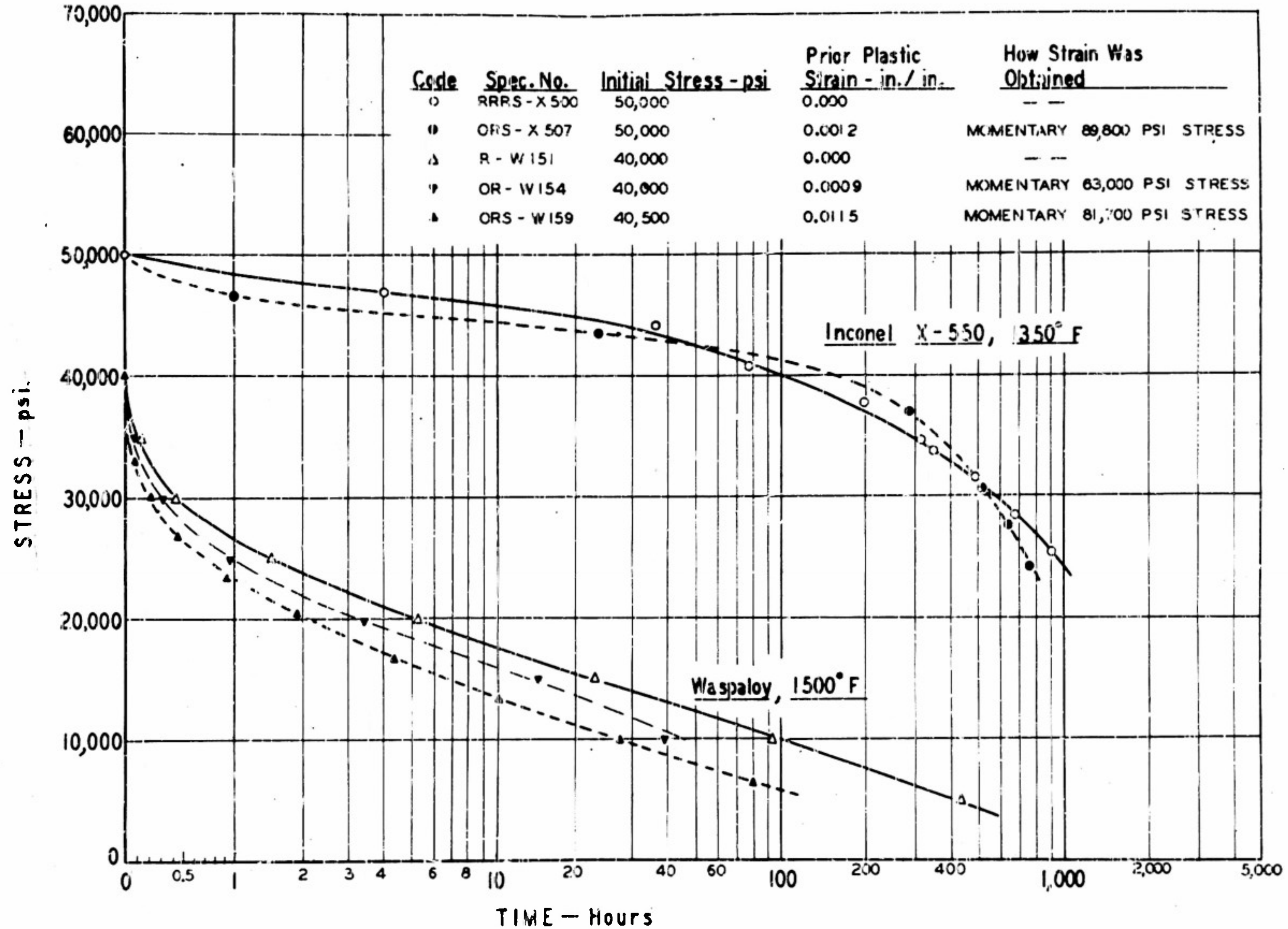


FIG. 27 EFFECT OF PRIOR PLASTIC STRAIN ON RELAXATION OF WASPALOY AT 1500°F AND OF INCONEL X-550 AT 1350°F

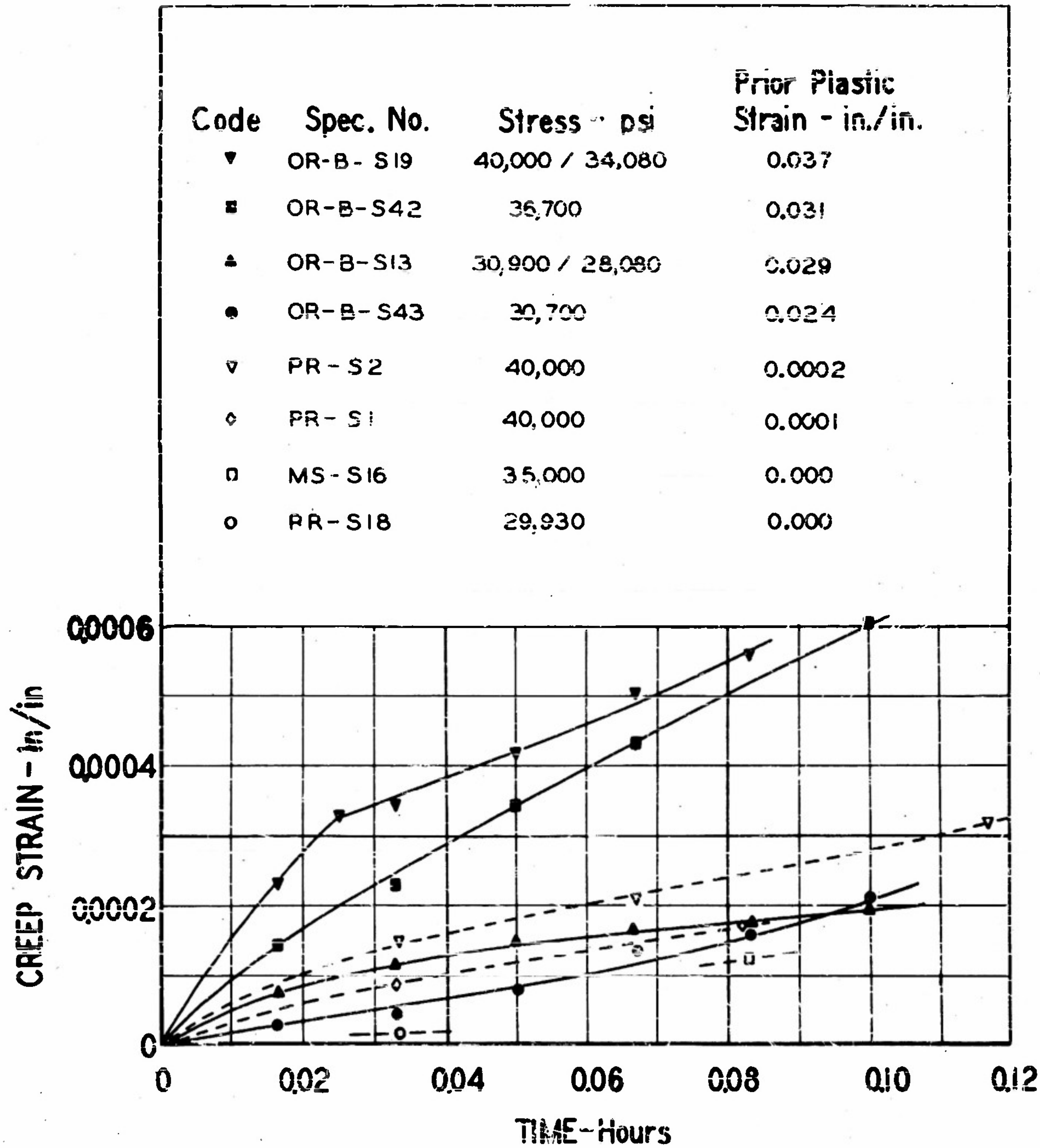


FIG.28 EFFECT OF PRIOR PLASTIC STRAIN FROM MOMENTARY OVERLOADING ON EARLY STAGES OF CREEP FOR S-816 AT 1350°F

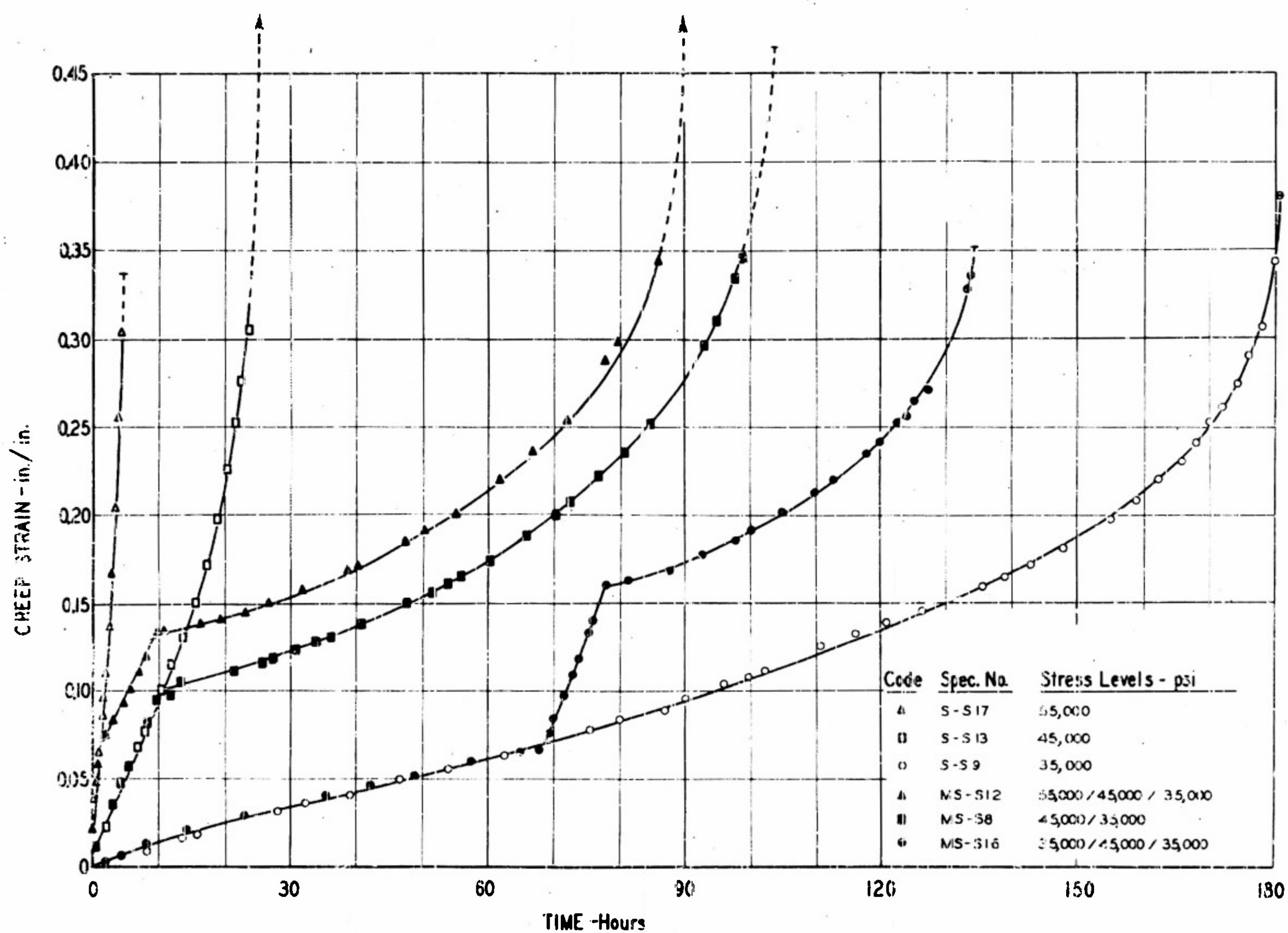


FIG.29 CREEP CURVES UNDER SINGLE- AND MULTIPLE - STRESS LOADING FOR S-816 AT 1350°F.

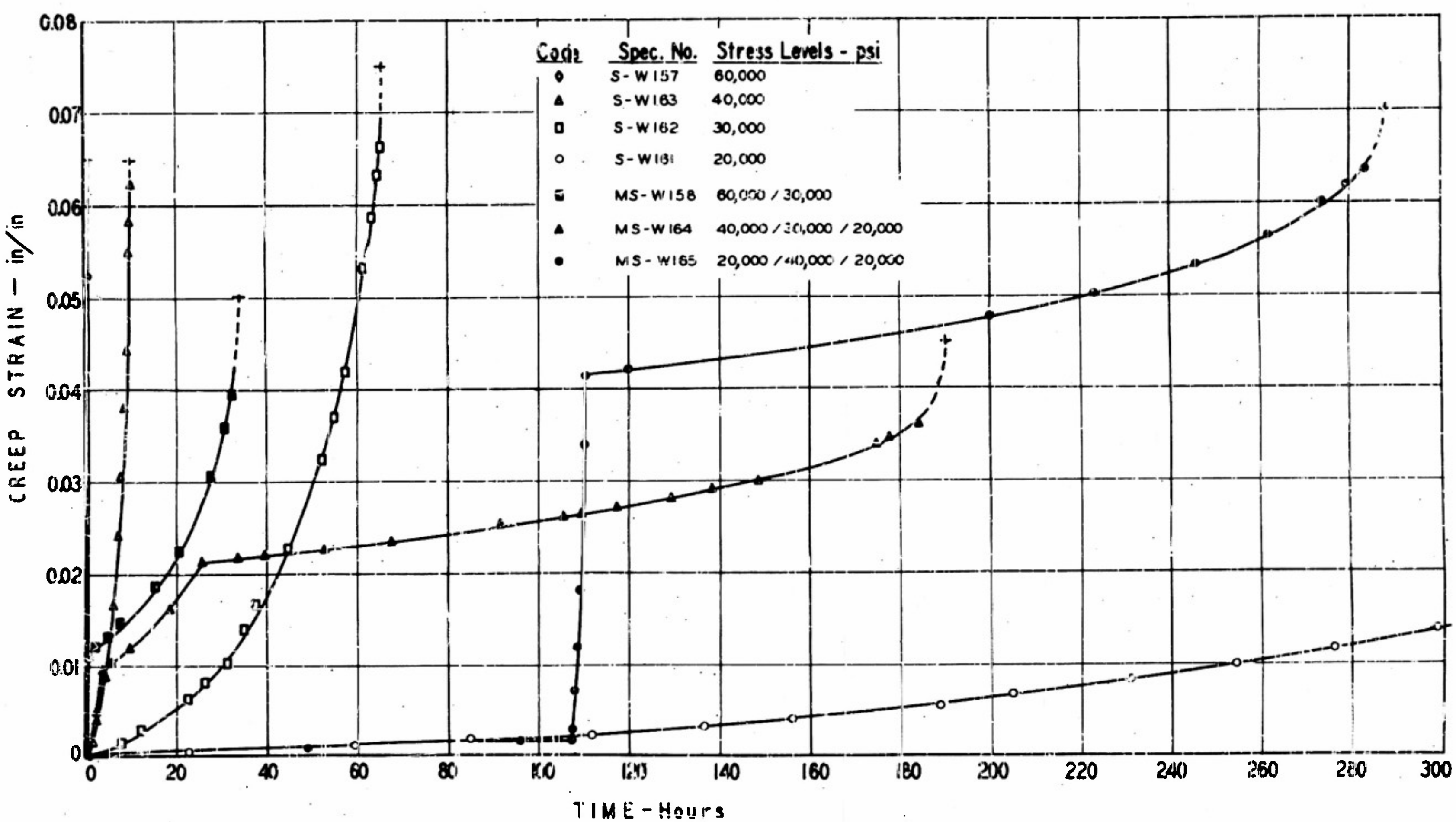


FIG. 30 CREEP CURVES UNDER SINGLE- AND MULTIPLE-STRESS LOADING FOR WASPALOY AT 1500 °F.

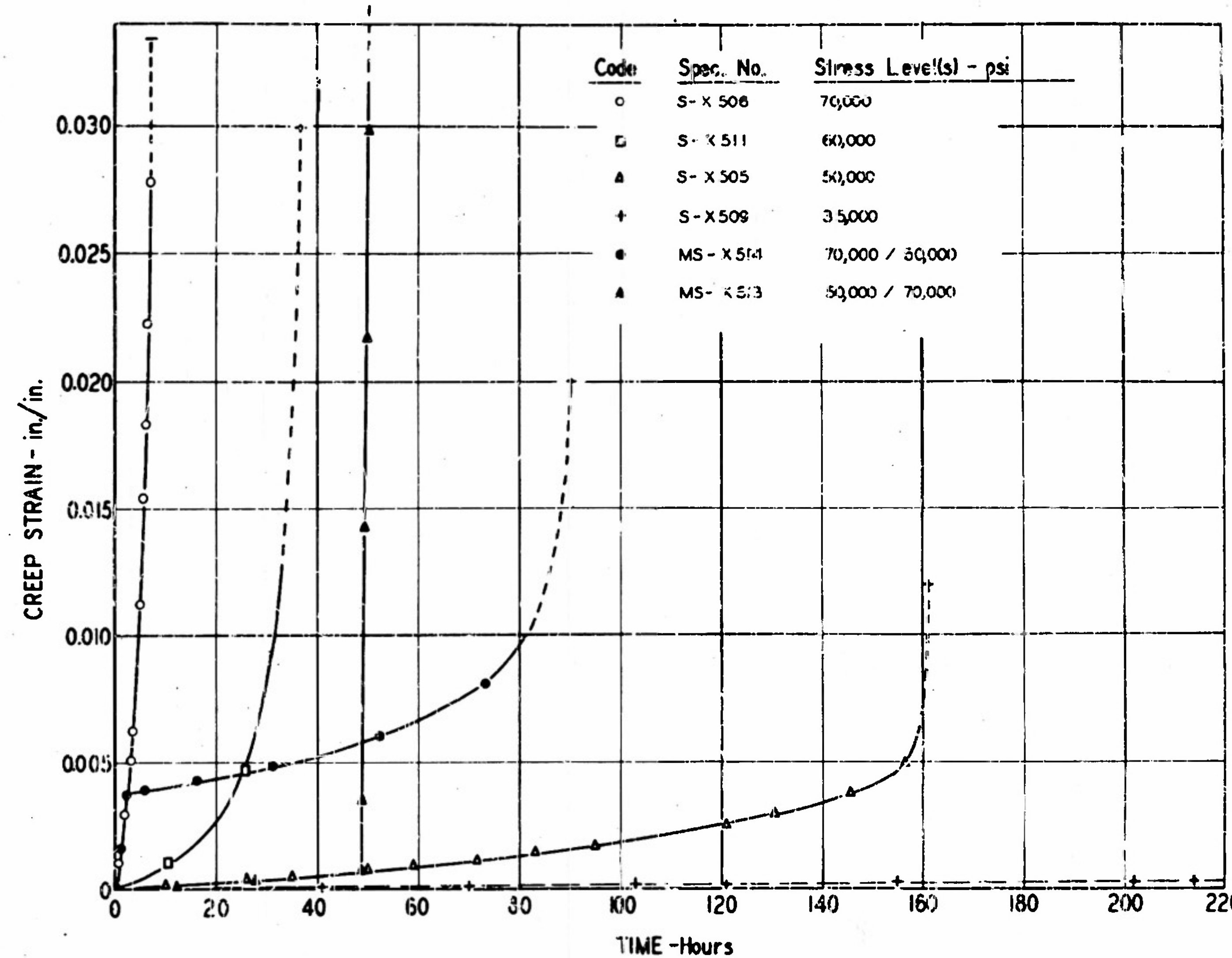


FIG. 31 CREEP CURVES UNDER SINGLE - AND MULTIPLE - STRESS LOADING FOR INCONEL X-550 AT 1350°F.

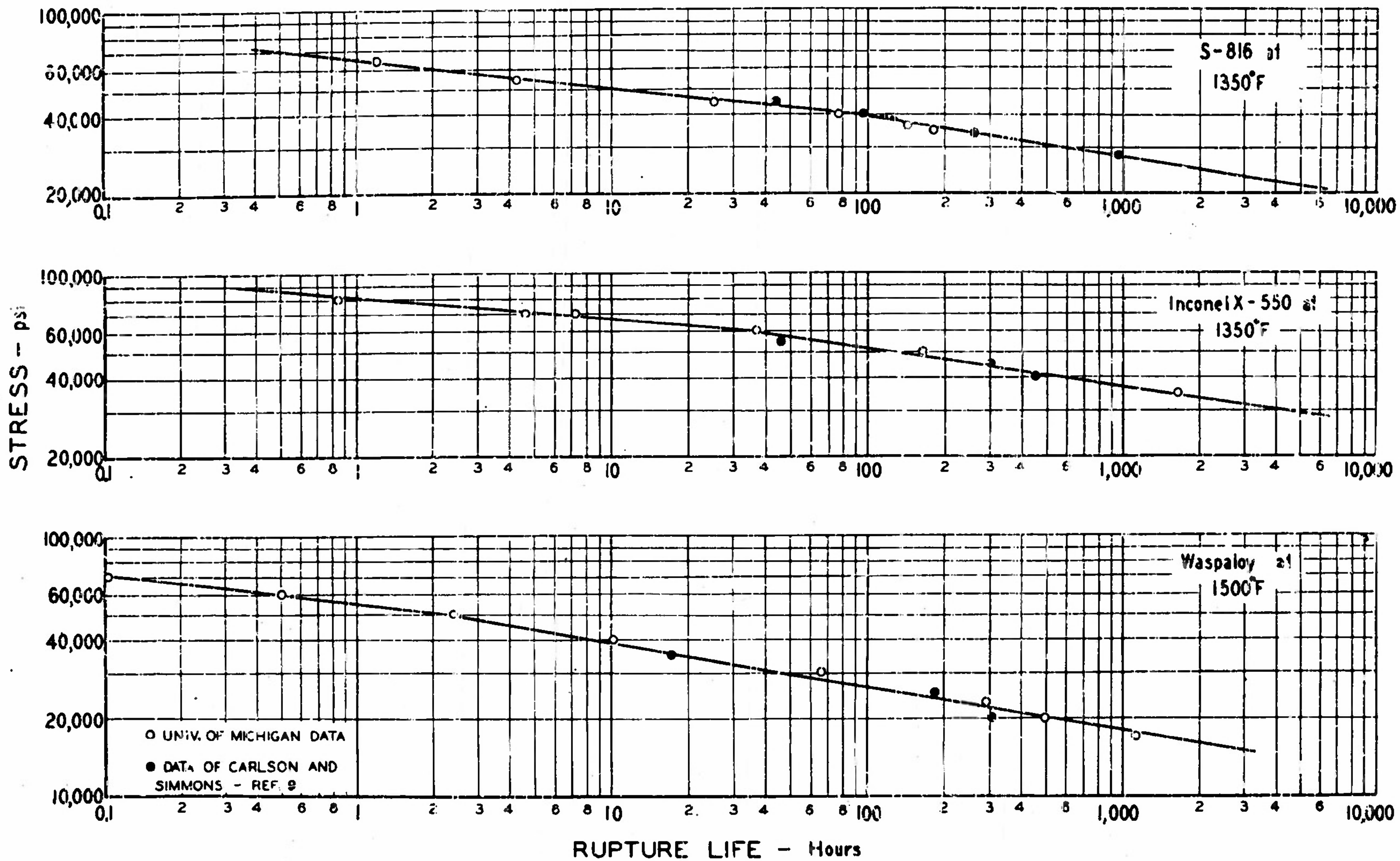


FIG. 32 STRESS VERSUS RUPTURE LIFE FOR THREE ALLOYS AT THE SINGLE TEMPERATURES STUDIED.

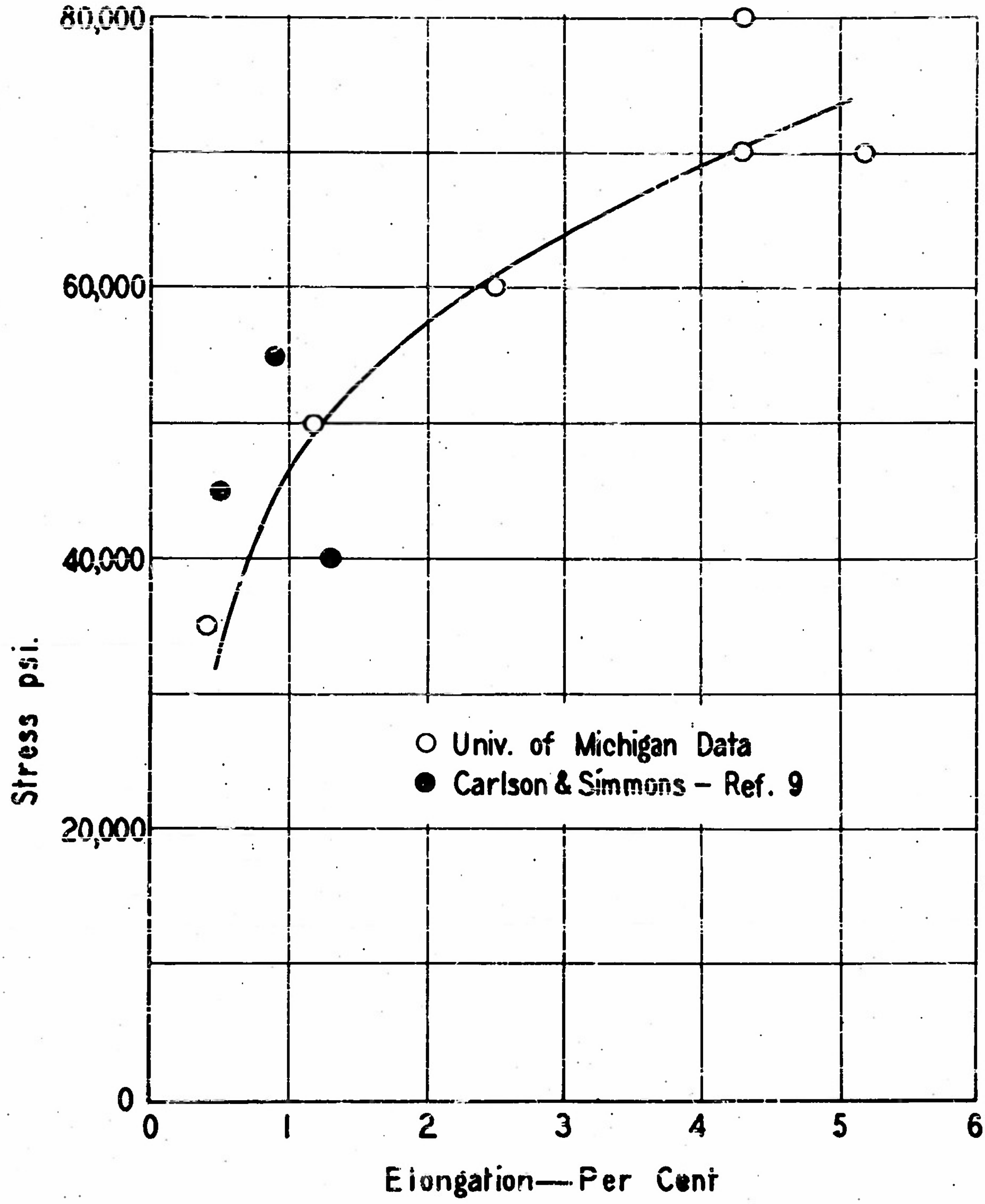


FIG.33 STRESS VS. ELONGATION FOR INCONEL X-550 AT 1350° F

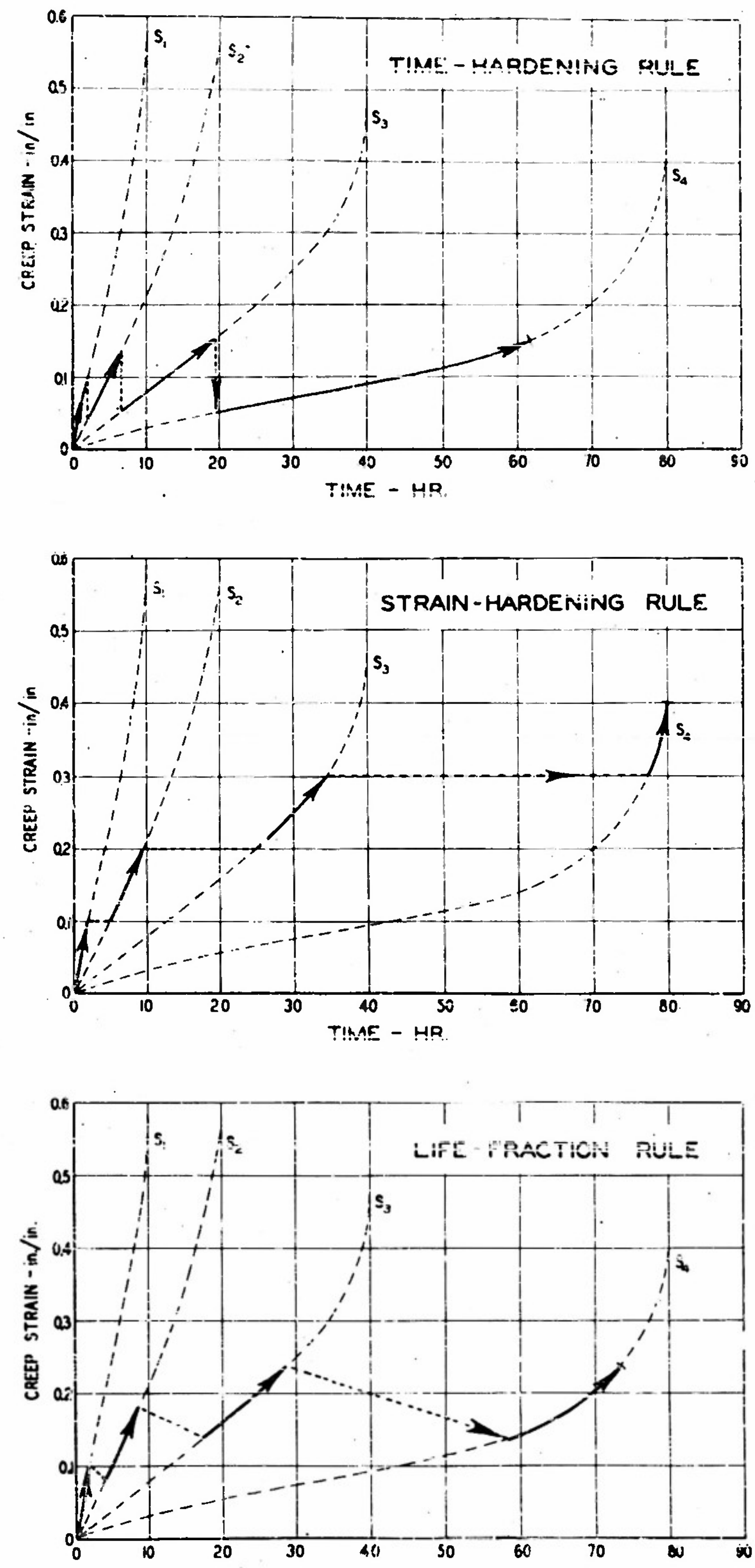


FIG.34 HYPOTHETICAL CURVES ILLUSTRATING RULES WHICH HAVE BEEN PROPOSED TO CORRELATE CREEP AND RELAXATION PROPERTIES.

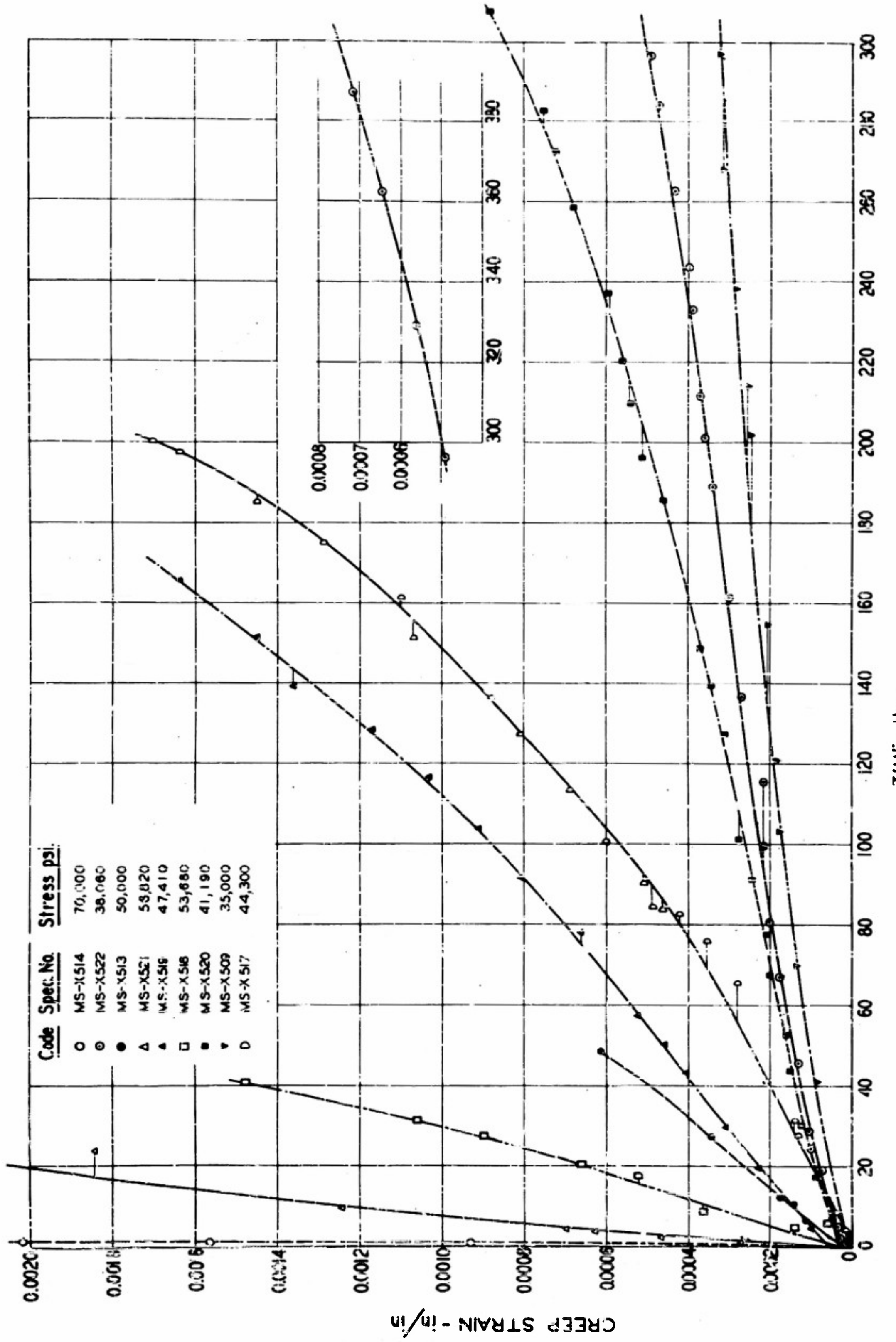


FIG.35 EARLY STAGES OF CREEP CURVES FOR INCONEL X-550 AT 1350° F

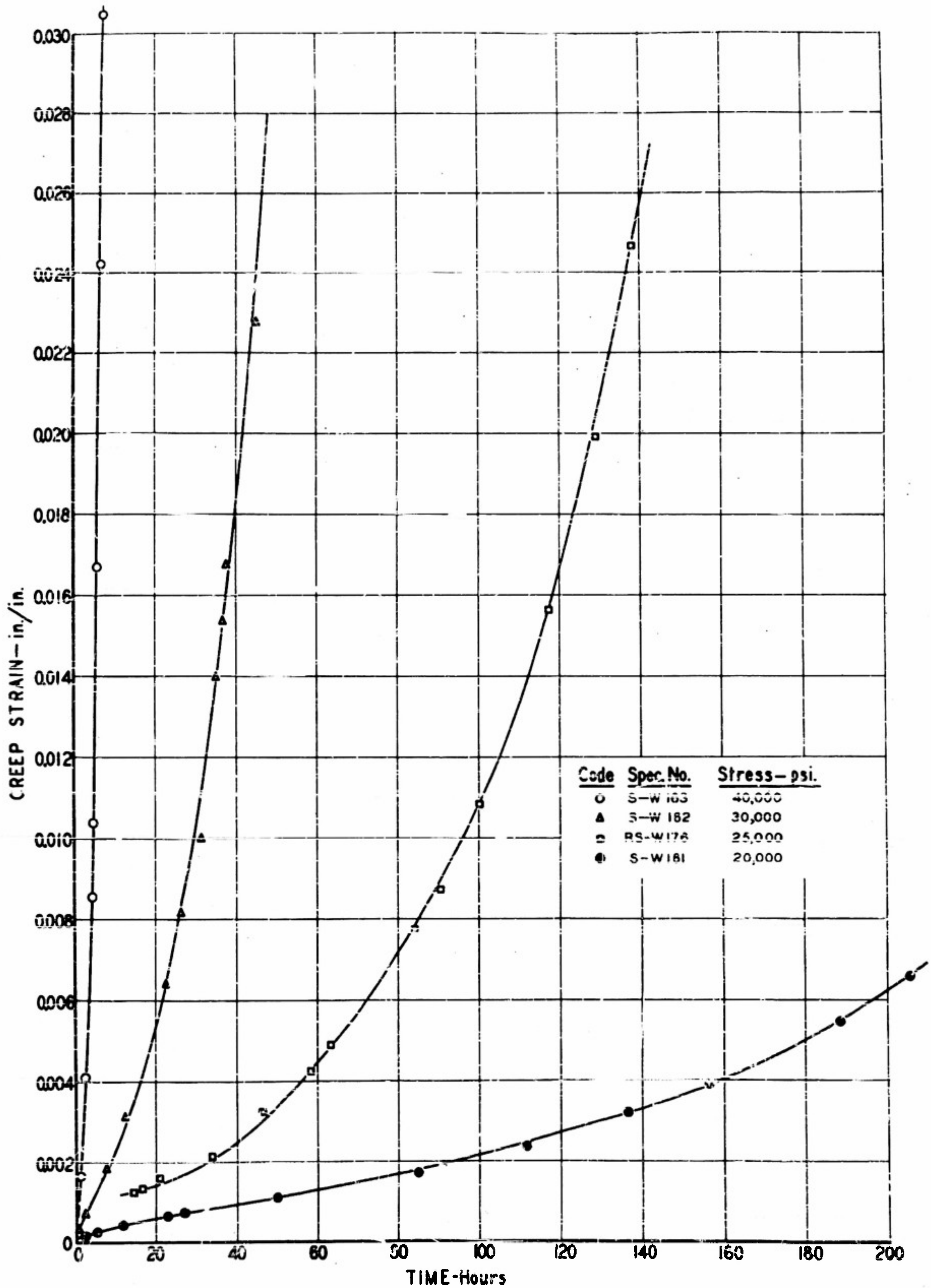


FIG.36 EARLY STAGES OF CREEP CURVES FOR WASPALOY AT 1500°F

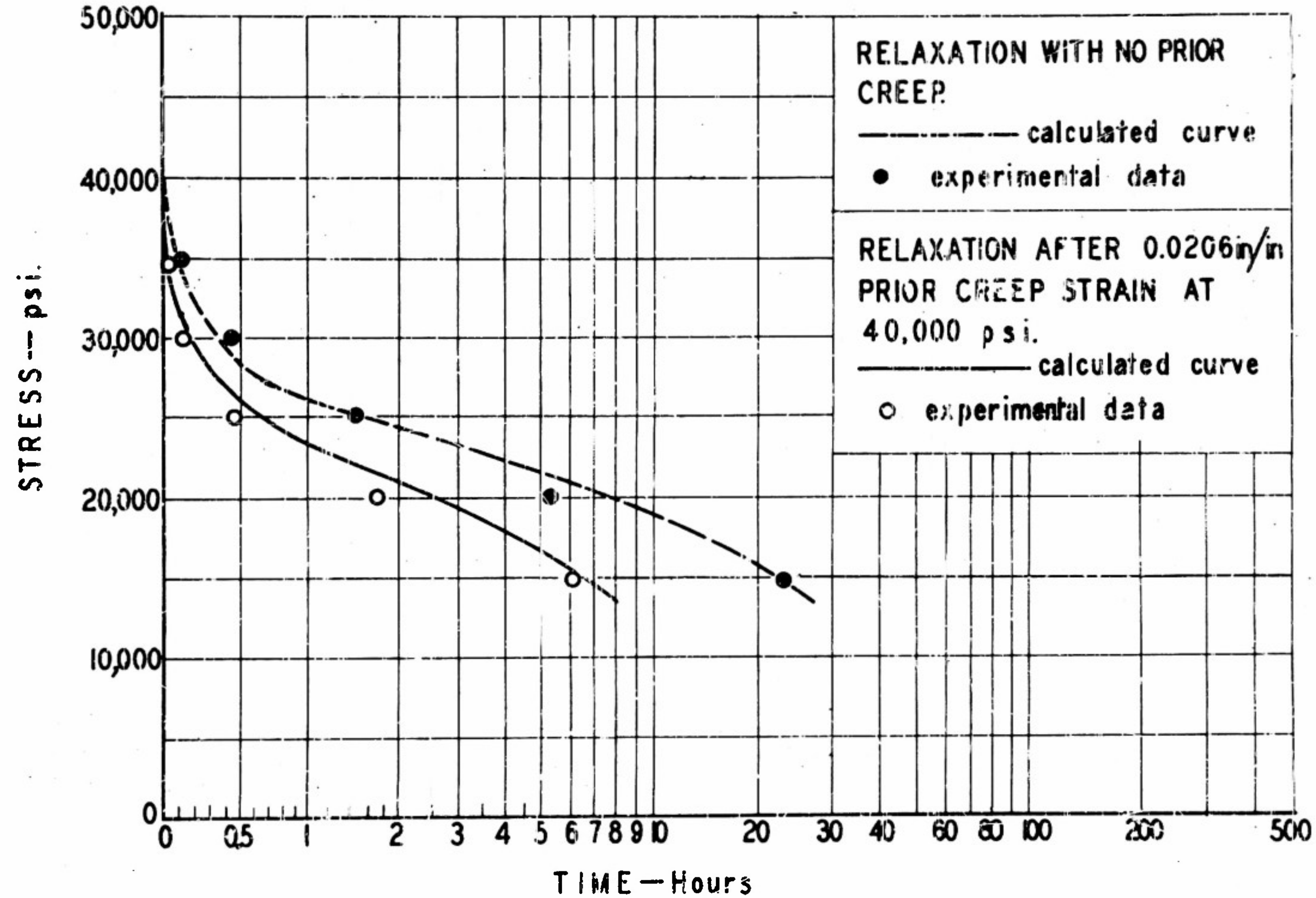


FIG.37 COMPARISON OF EXPERIMENTAL RELAXATION DATA FOR WASPALOY AT 40,000 PSI. AND 1500°F WITH RELAXATION CURVES PREDICTED FROM CREEP DATA.

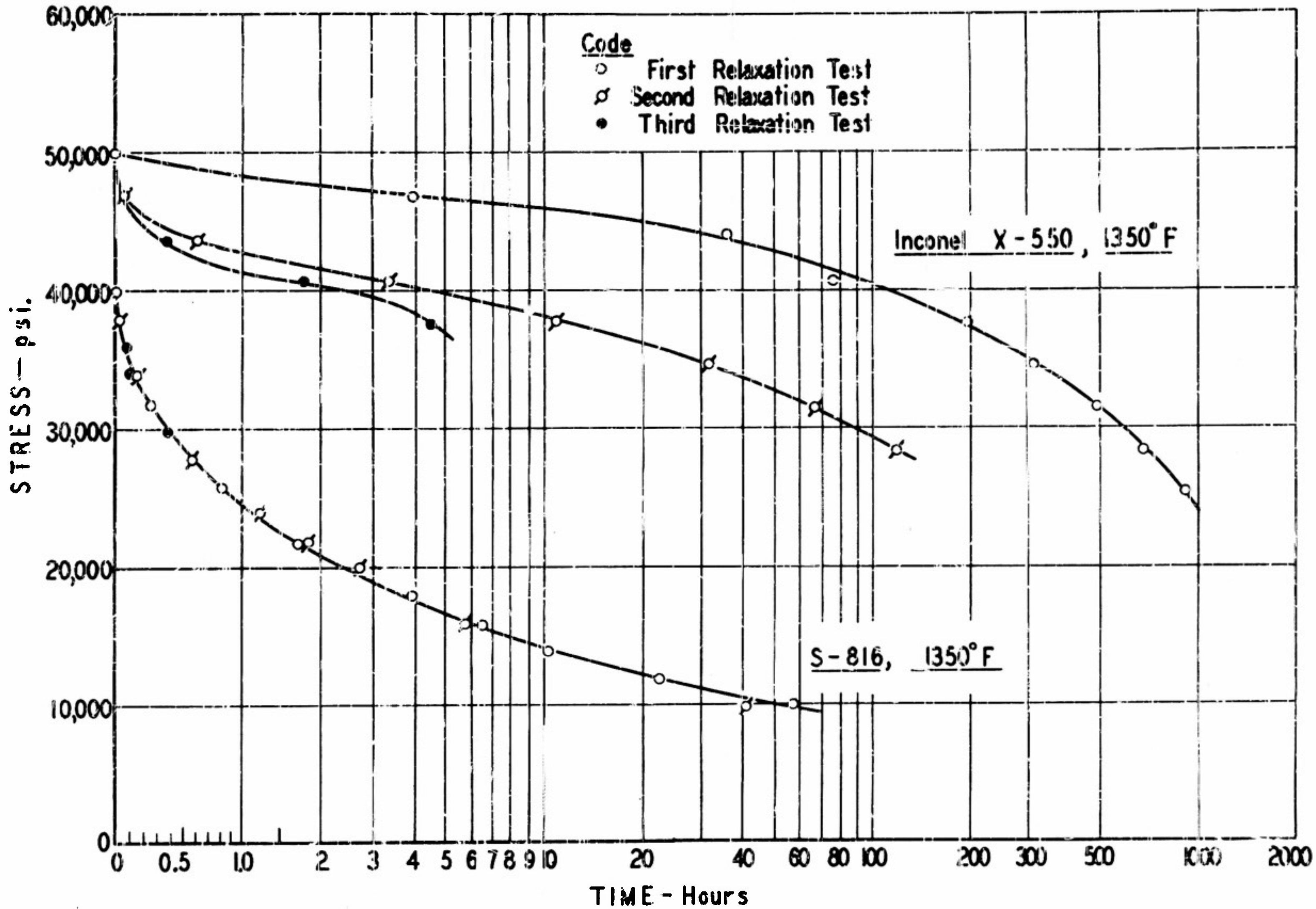
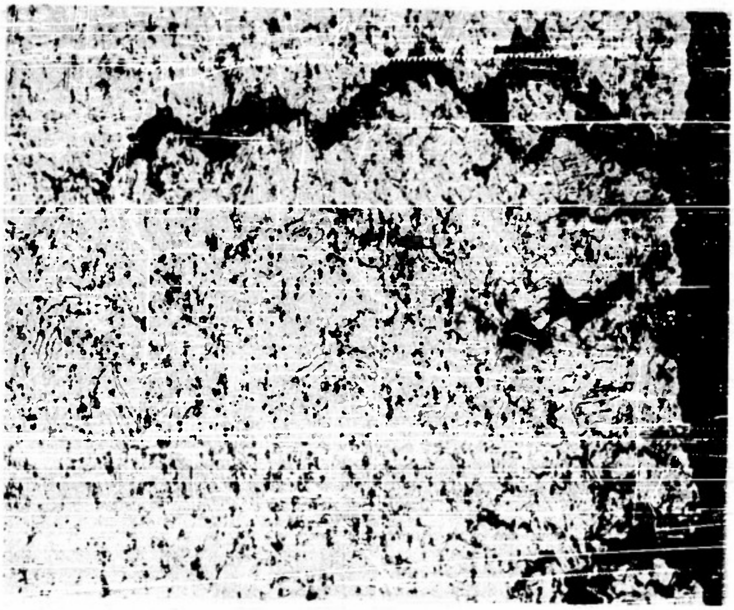
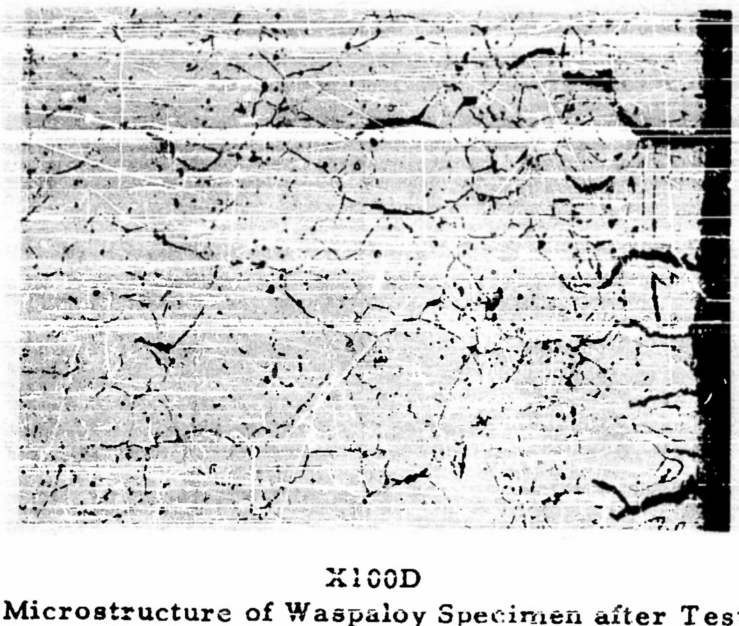


FIG.38 RELAXATION CHARACTERISTICS WHEN REPEATED TESTS WERE PERFORMED ON THE SAME SPECIMENS FOR S-816 AND FOR INCONEL X-550 AT 1350° F



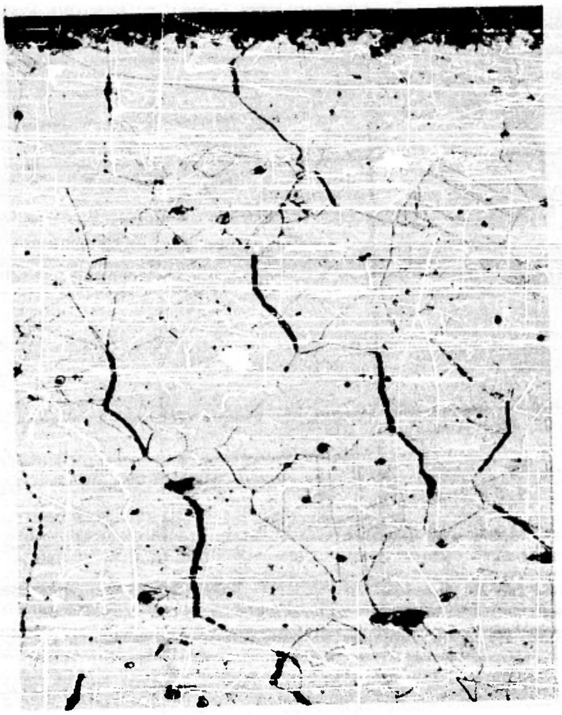
X100D

Figure 39. Microstructure of S-816 Specimen after Relaxation Plus Rupture Testing at 1350°F. (Specimen Number RRRS-S21. Relaxed 3 times from 40,000 psi; on the 3rd relaxation cycle, the stress was held constant when it reached 30,000 psi and fracture occurred in 669 hours. Total time at 1350°F was 766 hours.)



X100D

Figure 40. Microstructure of Waspaloy Specimen after Testing at 1500°F. (Fractured after 1129.8 hours under 17,000 psi.)



X100D

Figure 41. Microstructure of Inconel X-550 after Testing at 1350°F - Fractured after 1646.9 hours under 35,000 psi.

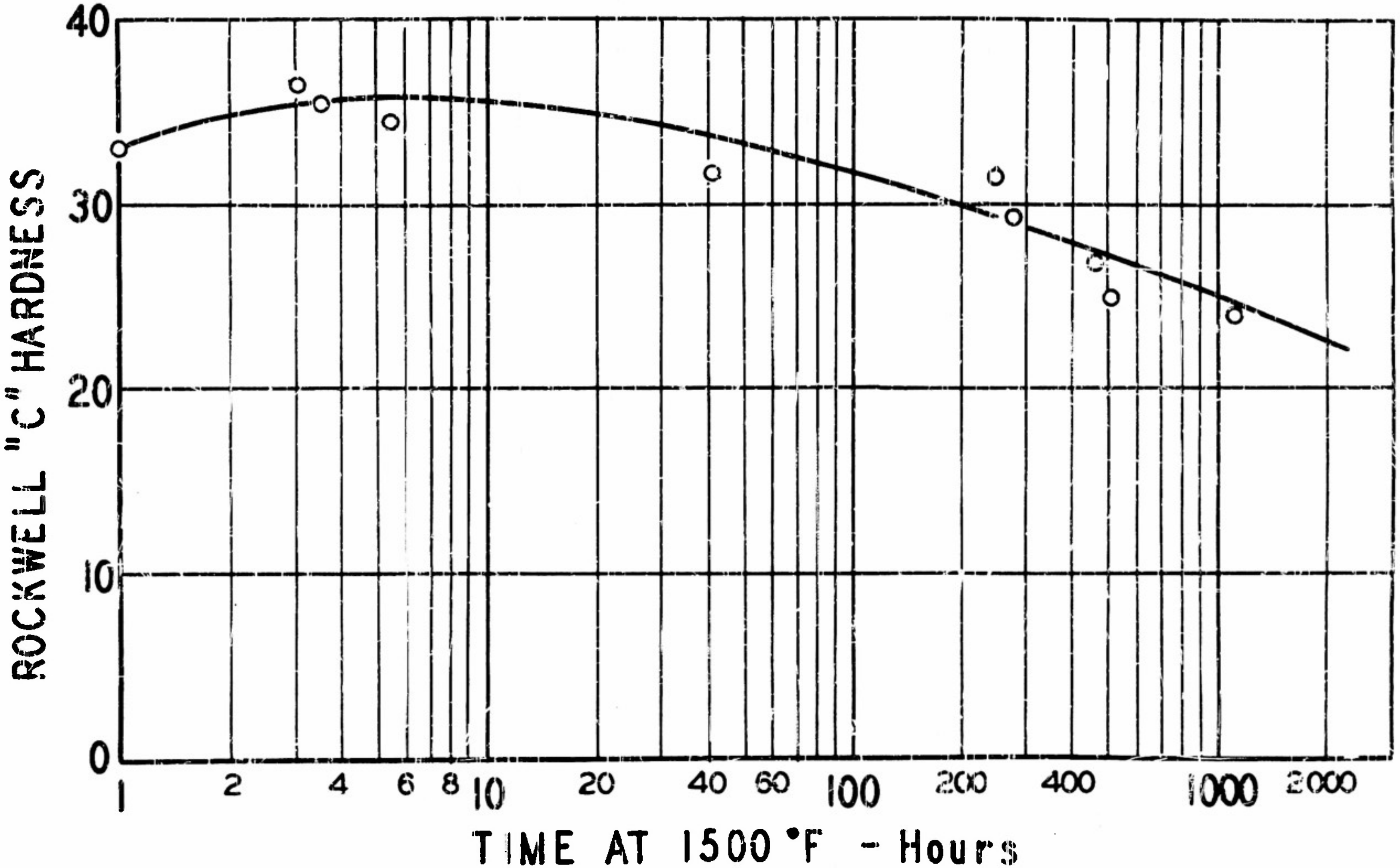
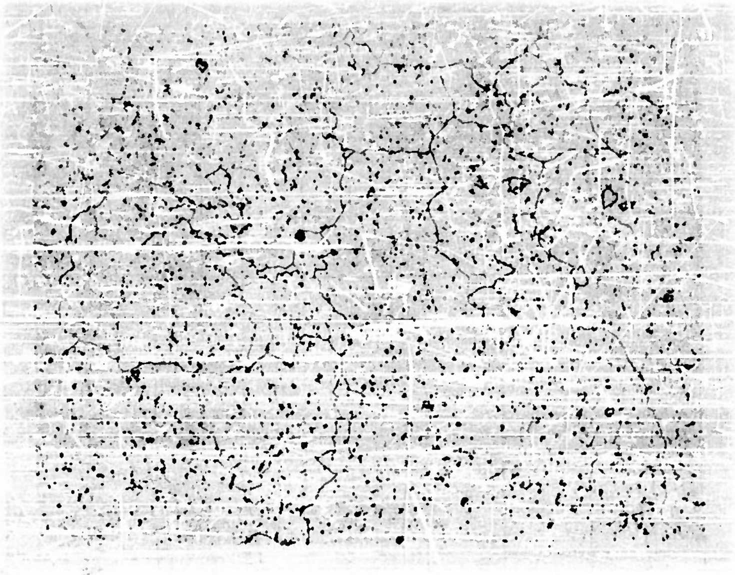
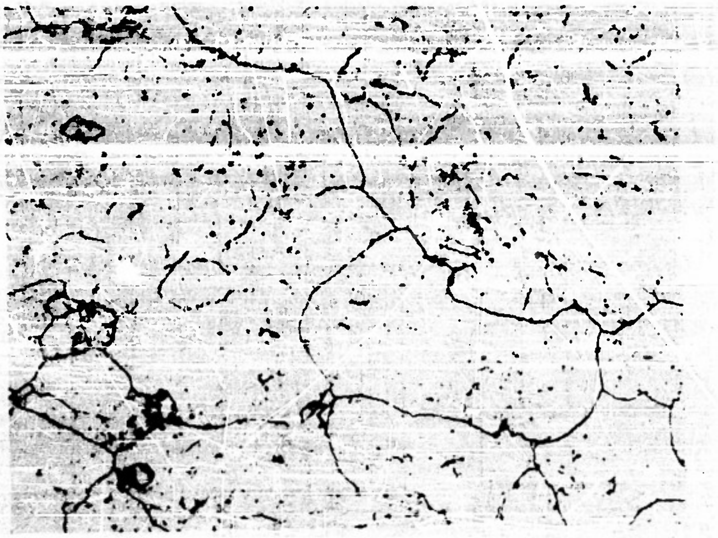


FIG.42 ROCKWELL "C" HARDNESS VERSUS TOTAL TEST TIME AT 1500°F FOR WASPALOY



X100D

Figure 43. Microstructure of S-816 Showing Abnormal Grain Growth. Reduced 1 percent by rolling at 75°F + 2150°F, 1 hour, water quench + 1400°F, 12 hours, air cool.



X100D

Figure 44. Microstructure of Waspaloy Showing Abnormal Grain Growth. Reduced 1-1/4 percent by rolling at 75°F + 1975°F, 4 hours, air cool + 1550°F, 4 hours, air cool + 1400°F, 16 hours, air cool.

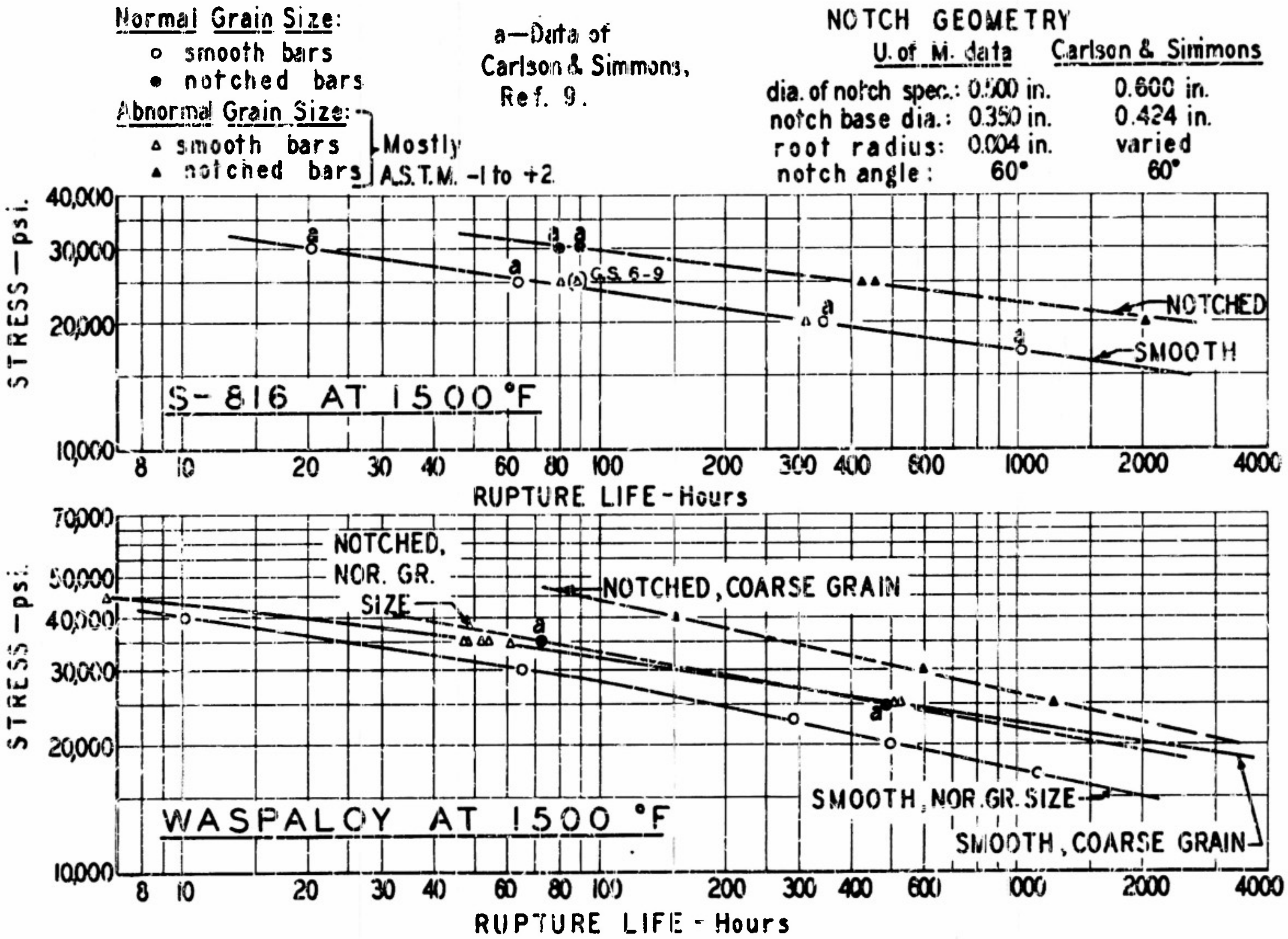


FIG. 45 EFFECT OF ABNORMAL GRAIN-SIZE RESPONSE ON RUPTURE LIFE AT 1500 °F FOR SMOOTH AND NOTCHED BARS OF S-816 AND OF WASPALOY.

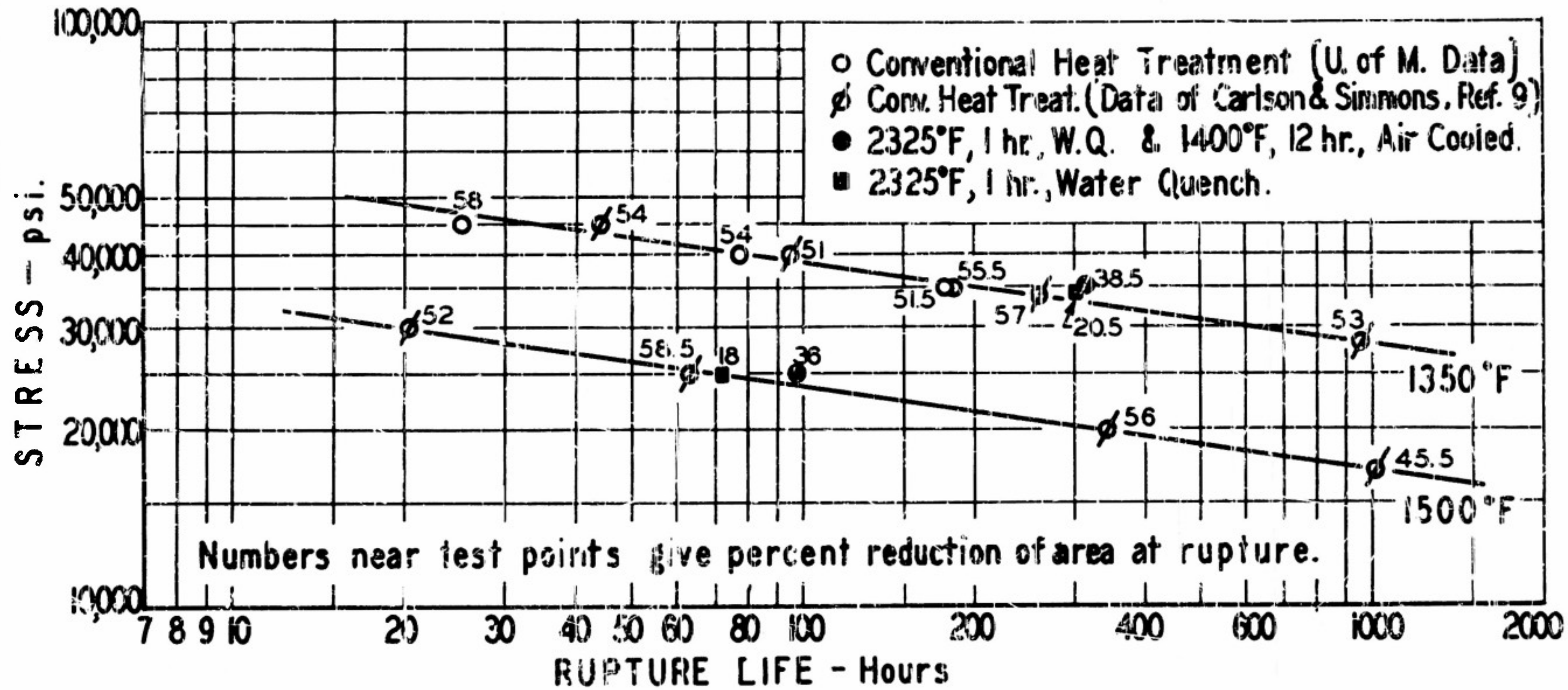


FIG. 46 SMOOTH-BAR STRESS-RUPTURE LIFE CURVES AT 1350°F AND 1500°F FOR S-816 WITH DEVIATION FROM CONVENTIONAL HEAT TREATMENT.

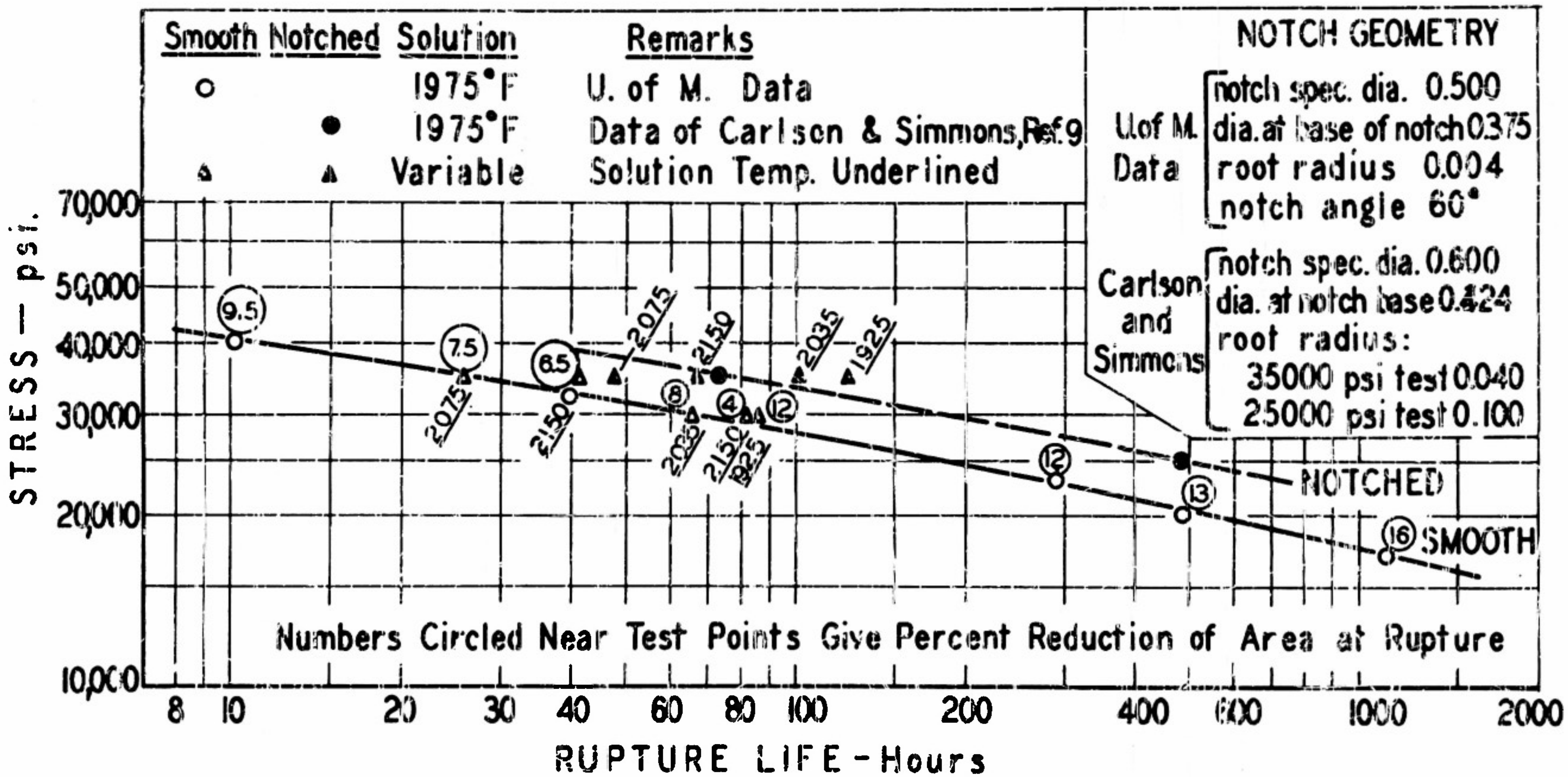
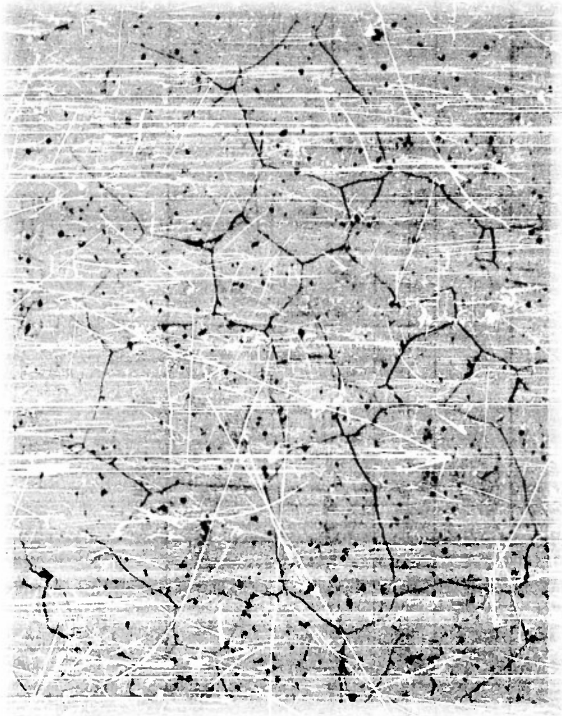


FIG. 47 EFFECT OF VARIABLE SOLUTION TEMPERATURE ON RUPTURE LIFE AT 1500°F OF SMOOTH AND NOTCHED BARS OF WASPALOY.



X100D

Figure 48. Original Microstructure of Waspaloy Solution Treated at 2150°F, 4 hours, Air Cool + 1550°F, 4 hours, Air Cool + 1400°F, 16 hours, Air Cool.

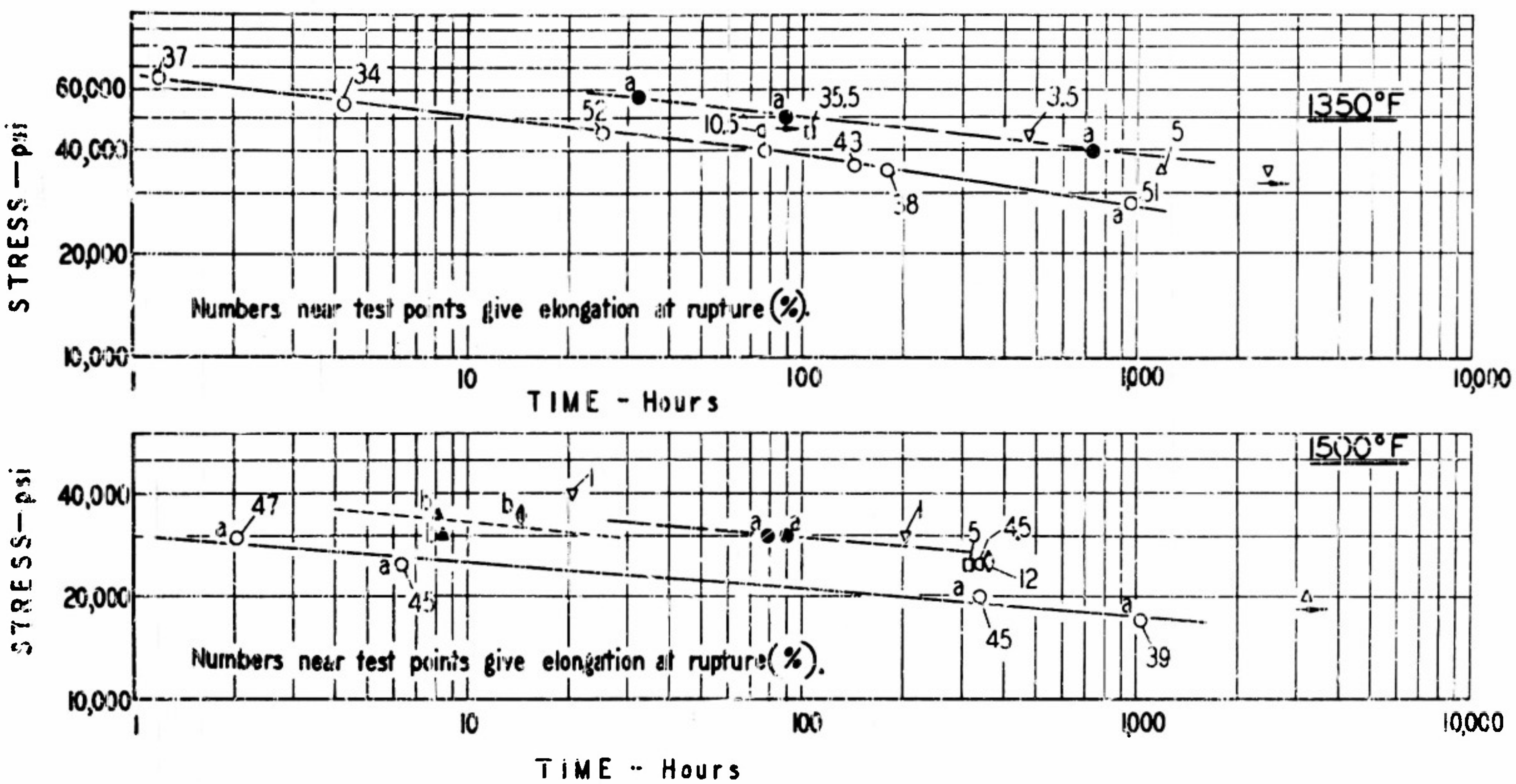
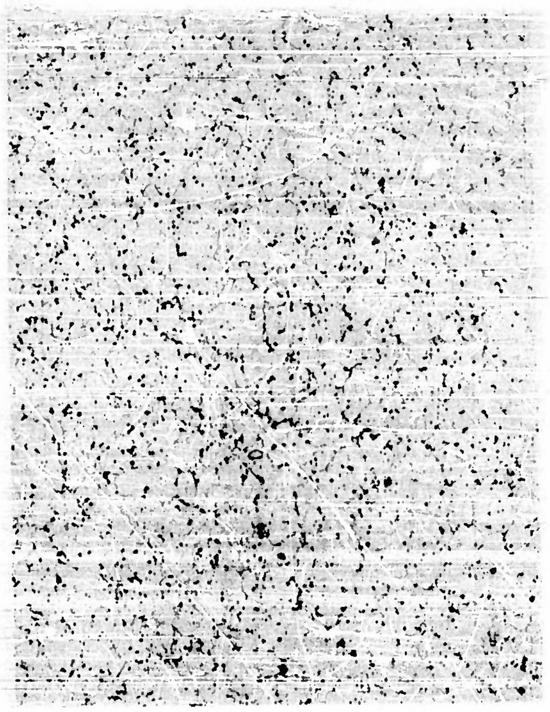


FIG.49 STRESS - RUPTURE TIME CURVES FOR S-816 AT 1350° AND 1500°F AFTER VARIOUS TREATMENTS.

SMOOTH	NOTCHED	HEAT TREATMENTS APPLIED
—o—	—●—	2150°F, 1 Hr., W.Q. & 1400°F, 12 Hr., A.C. (Conventional Treatment.)
□		2150°F, 1 Hr., W.Q. & 10 % Red. at 75°F & 1400°F, 12 Hr., A.C.
◁		2150°F, 1 Hr., W.Q. & 10 % Red. at 75°F
▽		2325°F, 1 Hr., W.Q. & 13.5 % Red. at 1200°F, A.C. & 1400°F, 12 Hr., A.C.
△	---▲---	2325°F, 1 Hr., W.Q. & 13.5 % Red. at 1200°F, A.C.
◊	---♦---	2325°F, 1 Hr., W.Q. & 5 % Red. at 75°F

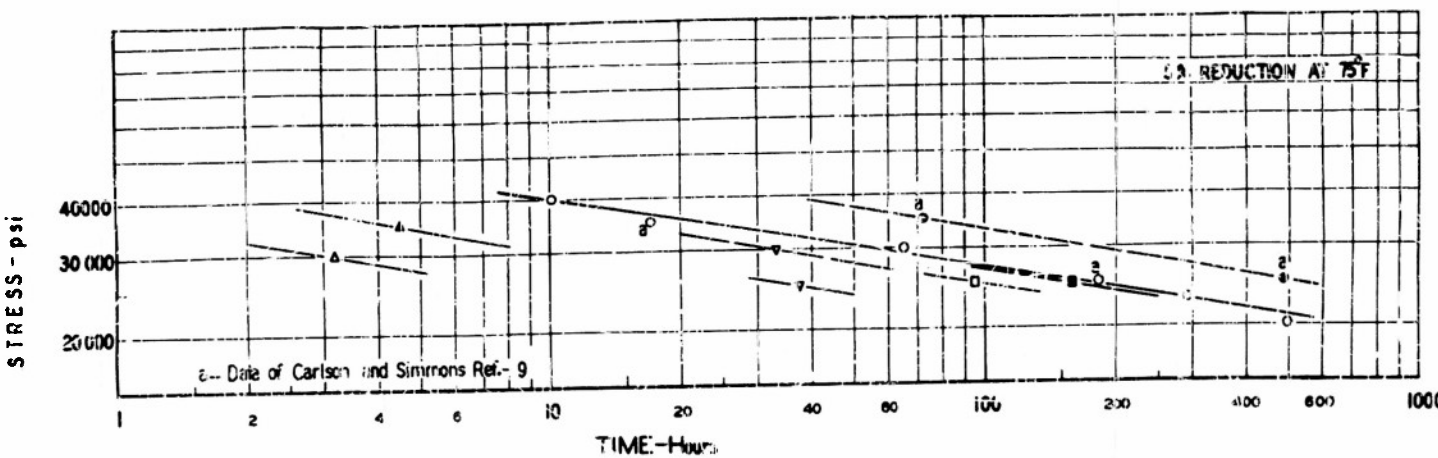
b—Notch Geometry:
D = 0.500 in.
d = 0.375 in.
r = 0.004 in.
60° ANGLE

a—Data of Carlson & Simmons, Ref. 9



X100D

Figure 50. Typical Original Photomicrograph of S-816 Rolled after Solution Treatment. (2325°F, 1 hour, water quench + 13.5 percent reduction at 1200°F + 1400°F, 12 hours, air cool.)



Code: Smooth	Code: Notched	Stage in Heat Treatment When Cold Rolled.
○	●	(Conventional heat treatment. Not rolled.)
△	▲	After solution; Before any aging.
▽	▼	After partial aging (1550°F, 4 hr).
□	■	After complete aging.

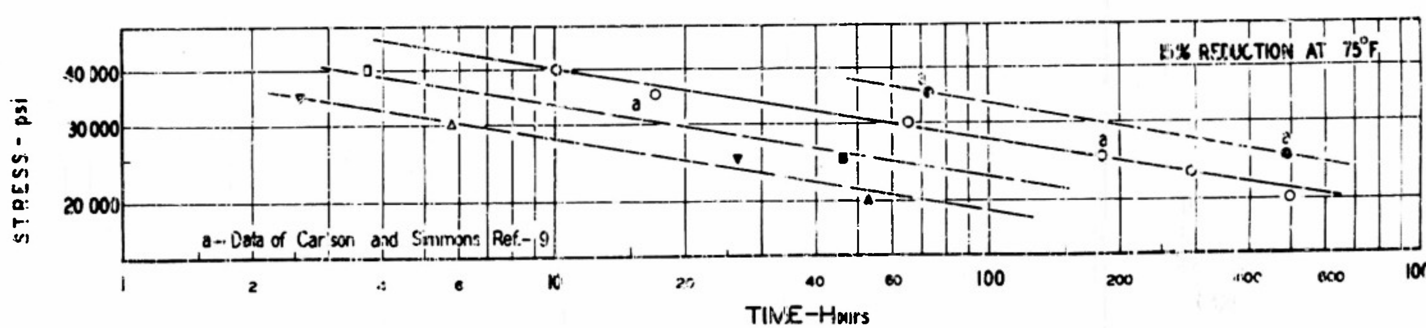


FIG. 51 EFFECT OF 5% AND 15% EXTRANEOUS COLD WORKING ON RUPTURE LIFE OF SMOOTH AND NOTCHED BARS OF WASPALOY AT 1500°F.

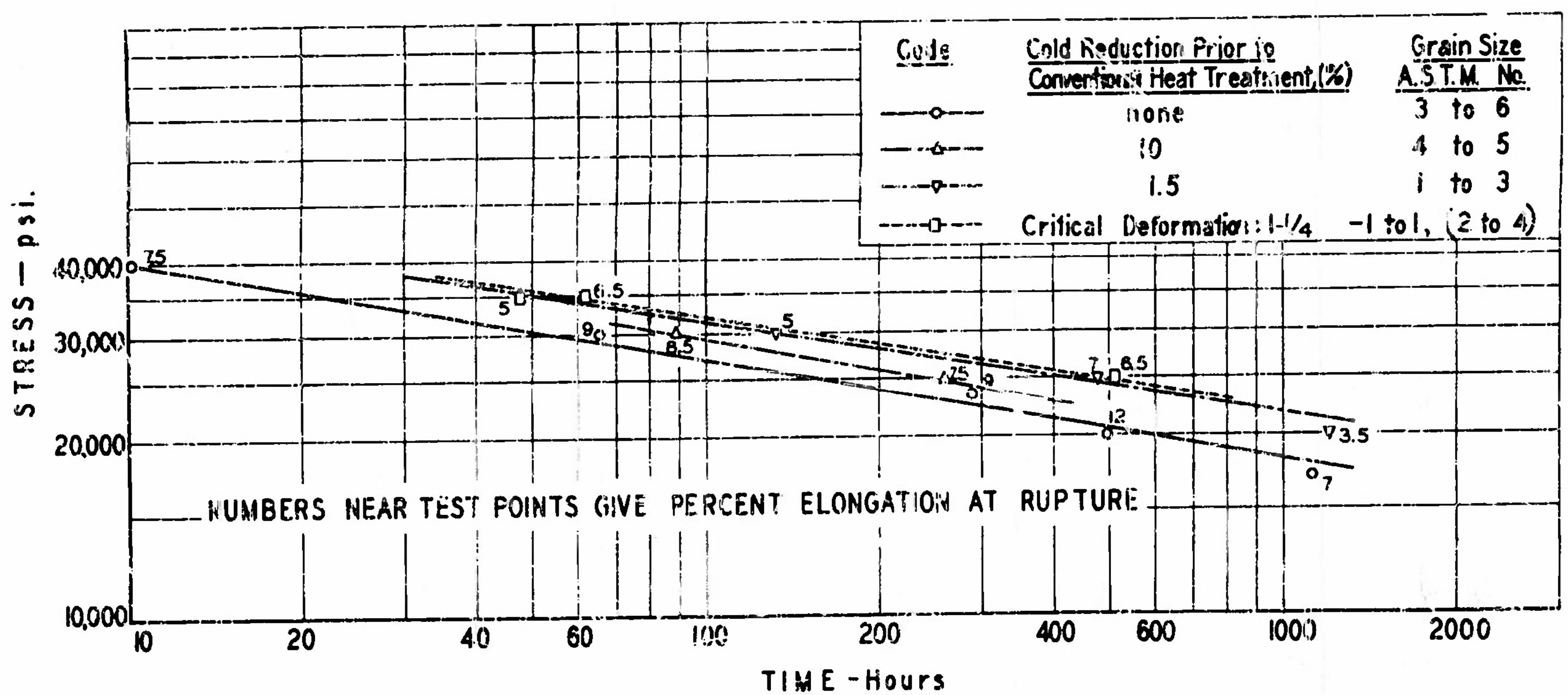


FIG. 52 SMOOTH-BAR RUPTURE PROPERTIES AT 1500°F FOR WASPALOY WITH DIFFERENT AMOUNTS OF COLD REDUCTION PRIOR TO CONVENTIONAL HEAT TREATMENT.

X100D

Figure 53. Typical Original Microstructure of Waspaloy after an Extraneous Cold Reduction during Processing. Solution treatment 1975°F, 4 hours, air cool + age 1550°F, 4 hours, air cool + reduced 5 percent at 75°F + age 1400°F, 16 hours, air cool.